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PHYSICO-METALLURGICAL ASPECTS OF WELDABILITY OF STEELS

1. Introduction

When evaluating weldability of steels we are going out from the requirement that welded joint is to be defect free and its utility properties should meet those required in the process of its exploitation.

Welded joint defect freedom is therefore the first quality requirement. Here, before all the absence of technological defects as lack of penetration, slag inclusions, pores and bubbles are meant.

To the most important defects belong cracks. Cracks or micro-cracks can arise in the process of welded joint fabrication /hot and cold cracks, micro hot tearing and lamellar cracking/ or during heat treatment of welded joints /annealing cracks/. Though, in principle, it is possible to determine the welding conditions under which no cracks or micro-cracks occur /pre-heating, interpass temperature, filler material, specific heat input.../, such process would not be economically viable.

For this reason during the last years, on the basis of linear fracture mechanics knowledge the presence of such defects in weld metal is admissible dimensions of which are sub-critical and which - with respect to the steel's properties-do not ^{represent} vary danger of defect propagation up to \leq the critical size a_c in the process of welded joint exploitation.

As regards the requirements on welded joint properties, they can be widely varied depending on the medium in which

the welded joint will operate. E.g. for dynamically loaded constructions be most important are fatigue properties including lowcyclic fatigue. Steels work^{ing} in corrosion medium must show very good corrosion resistance of welded joints. Steels used in power industry must have good creep properties of welded joints and good structural stability.

Of course, the welded joint properties will depend on physico-metallurgical changes occurring in the welded joint zone and on composition, production method, and heat treatment of steel alternating these changes.

To these physico-metallurgical changes the following belong:

- structural changes connected with polymorphy,
- substructural changes,
- precipitation changes,
- deformation processes /change in configuration and dislocation density/,
- changes in impurities distribution /sulphides, oxides.../.

The effect of chemical composition on weldability of steels

It is impossible to present a generally valid analysis of the chemical composition effect, as the effect of individual elements can in various steel types be different. Nitrogen, for example. In case of mild steels its effect is very negative, as it increases the steel's susceptibility to aging. There exist, however, a whole group of steels /with vanadium or aluminium, the so-called "IN process"/ in which the nitrogen content is deliberately increased.

Similarly, in high strength and particularly refined steels the effect of nitrogen is not dominant.

Generalizing, it can be said that the effect of chemical composition on weldability of steels can be expressed by parametric relations:

The best known of them is the carbon equivalent C_{eq} /1/ relation:

$$C_{eq} = C + \frac{Mn}{6} + \frac{Cr+Mo+V}{5} + \frac{Ni+Cu}{15}$$

This formula takes into account also the effect of base chemical composition of steel, while it is supposed that the content of impurities and gases in the steel is kept within a certain tolerance. For plate thicknesses up to 20 mm and carbon equivalent values up to 0.40% no limiting conditions of weldability of steels are necessary, i.e. neither pre-heating nor limiting of heat input volume are necessary. It is probable that for carbon equivalent values ranging from 0.40 to 0.60% the pre-heating temperatures up to 150°C should be used. The carbon equivalent value indicates the level of transformation embrittlement in the heat affected /under-bead/ zone. The V_1 critical cooling rate of 300 °C at which in the under-bead zone purely martensitic structure forms can according to Maynier /2/ be expressed as follows: $\log V_1 = 3.00 - /4.620C + 1.05 Mn + 0.54Ni + 0.50Cr + 0.66Mo/ \text{ } ^\circ\text{C/s}$, where C, Mn ... are weight percentages of individual elements in the steel.

If actual cooling rate V_R is lower than critical one heterogeneous mixture of disintegration phases creates in the under-bead zone.

According to the welding method and parameters two types of disintegration structure can occur in heat affected zone viz. ferrite-pearlitic and bainitic.

Ito e.a. /3/ tried to characterize the contribution of individual elements to hardening of under-bead zone with both kinds of reaction. To this purpose the D_{IC} and D_{IB} hardening indices for both structure types were introduced. It was found that at ferrite-pearlitic structure the D_{IC} hardening index depends on grain size. The coarser is grain the higher is D_{IC} index, as it depends on carbon content only. E.g. for grain size 1 according to ASTM $D_{IC} = 0.548xC$ and for grain size 10 it is only $0.266xC$. If the under-bead zone structure is formed by the upper bainite, then the $D_{IC} = 0.494xC$ and does not depend on grain size.

Consequently, in case of mild un-alloyed steels, where either ferrite-pearlitic or upper bainitic structure can occur in the under-bead zone, hardening of zone depends exclusively on carbon content and steel's susceptibility to primary grains growth.

When the lower-bainitic reaction prevails, the hardening index according to /3/ can be expressed as follows:

$$D_{IB} = 0.5 \%C /1+0,64\%Si/ /1+4,10\%Mn/ /1+2,83\%P/ /1-0,62\%S/ \\ \cdot /1+2,33\%Cr/ /1+0,52\%Ni/ /1+3,14\%Mo/ /1+0,27\%Cu/ \\ \cdot /1+1,5 /0,90-1\%C/ \%B/ .$$

As evident from the above expression, at bainitic reaction the HAZ hardening is affected by a whole group of steel's chemical composition elements. The D_{IC} and D_{IB} hardening indices can be taken as alternatives of carbon equivalents: they indicate weldability of steel indirectly only.

Ito e.a. /3/ introduced therefore for structural changes in under-bead zone the so called corrected P_{BA} and P_{BB} expressions.

1. If in this zone martensite and lower bainite prevail, then

$$P_{BA} = C - \frac{1}{6} Cu - \frac{1}{5} Ni - \frac{1}{11} Mo + 4S$$

2. If upper bainite and ferrite are created, then

$$P_{BB} = C + \frac{1}{10} Mn + \frac{1}{6} Cu + \frac{1}{12} Cr + \frac{1}{13} Mo - \frac{1}{50} Ni$$

Now, the relation between P_{BA} or P_{BB} and the expected transition notch toughness temperature /ChV/ can be obtained:

Transition temperature /°C/	-40	-20	± 0	+20	+40
P_{BA} value	-0,2	-0,08	+0,07	+0,20	+0,32
P_{BB} value	+0,12	+0,18	+0,23	+0,30	+0,37

For the group of refined high strength steels Sato e.a. /4/ introduced the carbon equivalent values representing notch toughness /ChV/ at 0°C / $C_{eq. /BvEo/}$ / and at -40 °C / $C_{eq. /BvE-40/}$ /

$$C_{eq. (BvEo)} = \%C + \frac{1}{53} \%Si + \frac{10}{85} \%Mn + \frac{1}{36} \%Ni - \frac{1}{210} \%Cr + \frac{1}{17} \%Mo - \frac{2}{3} \%V + \frac{10}{13} \%Al + 5,1\%B,$$

$$C_{eq. (BvE-40)} = \%C - \frac{10}{74} \%Si + \frac{10}{92} \%Mn - \frac{1}{173} \%Ni + \frac{1}{109} \%Cr - \frac{1}{118} \%Mo - \frac{10}{14} \%V + \frac{2}{15} \%Al + 1,7\%B$$

Ito e.a. /3/ expressed the transition temperature /ChV/ in submerged-arc welded joints with heat input of 55 kJ/cm as follows:

$$vT_{rs} / ^\circ\text{C} / = -70 + 290(\%C) + 28(\%Mn) + 46(\%Cu) + 25(\%Cr) + 23(\%Mo) - 6(\%Ni)$$

Of course, the validity of all above given relations is limited to those steels or welding technologies for which they were derived. Anyway, with their help the effect of chemical composition of steel on its weldability can be determined.

The effect of steel production method on its weldability

Under the term "steel production method" the type and capacity of furnace in which steel is melted, way and kinetics of deoxidation and weight of ingots are understood. Way of melting can affect the content of impurities and gases in the steel.

The deoxidation kinetics influences to a great extent the quality of sulphide inclusions, particularly in case of Ti and Nb micro-alloyed steels.

In construction steels the major content on sulphur is bound to manganese. The melting temperature of γ -MnS is higher than the temperature of steel solidus, sulphidic inclusions do not therefore dissolve in heat affected zone even at higher affecting temperatures /Fig. 1/. However, if the inclusions are partially bound in form of titanium or niobium sulphides or if MnS is alloyed with some further elements,

then the melting temperature of sulphide can drop deep below the solidus temperature /5/. Such sulphide inclusions dissolve in a part of HAZ and sulphur concentrates in primary grains boundary zones /Fig. 2/. In this way at the temperatures above the sulphide melting point thin film of melted metal creates, what decreases the cohesion between individual grains. Under tensile stress the grains can separate, i.e. micro hot tearing can occur, or the unhomogeneities can ^{the} on one hand propagate to weld metal and on the other hand to a part of HAZ in form of lamellar fissuring. Low sulphides melting point can result in formation of cracks in welded joint. The unfavourable effect of sulphides is observed not only in case of micro-alloyed steels but everywhere where the AlTi combination is used as the last stage of melt deoxidation. This combination can be particularly dangerous for the welded joint quality in case of alloyed /e.g. 3.5 % Ni/ steels where melt solidifies as austenite, rightly of peritectic reaction. This kind of solidification prefers dendritic segregation. Though the steel deoxidation by AlTi combination is very advantageous from the metallurgical viewpoint, it may unfavourably affect the weldability of steel.

In production of not only micro-alloyed steels plates ever more often uni-directional rolling is used. In this case the deformable sulphides elongate to fibres long up to 1 or 2 mm, this unfavourably affecting not only the properties of plate in cross direction but also the steel's susceptibility to lamellar tearing at welding. To transform the manganese sulphides

to globular oxysulphides which are more difficult to roll it is suitable to add to the mould or pan small amount of any element from the group of lanthanides or zirconium.

It has been mentioned already how important is the sulphides inclusions melting point for the evaluation of weldability of steel. The disadvantage of niobium or titanium sulphides is that when they are melted and concentrate in the primary grains boundary zone the concentration of alloying elements in this zone increases and, consequently, during cooling down /solidification/ these formations precipitate not only as dendritic sulphides but also as niobium or titanium sulpho-nitrides. As the niobium or titanium nitrides or carbo-nitrides precipitate as a thin film of thickness often less than 100 Å, the unfavourable effect of sulphide to cohesion strength of grain boundaries is multiplied. Typical morphology of niobium sulpho-nitride in the over-heated zone of HAZ is shown in Figs 3a and 3b.

3. The effect of rolling and heat treatment manner on the weldability of steel

In the production of plates of fine-grained and particularly micro-alloyed steels special rolling and cooling procedures are often used, these being referred to as controlled rolling or controlled cooling. The aim of such a procedure is to obtain the optimum properties of plates. When Nb, V or Ti are used as micro-alloying elements, the controlled rolling enables to decrease carbon content in steel and, through its partial binding to the alloying elements, to decrease also

the content of pearlite. In accordance with it there exist steels with reduced pearlite content and even pearlite-free steels. Weldability of such steels is good also due to the fact that their heat affected zone width is smaller than that of conventional steels.

The same is valid also for heat treatment. Heat treatment irrespective whether homogenization, normalization or stress relieving are concerned, homogenize the properties of plates. The application of heat treatment in the temperature range A_1-A_3 improves e.g. weldability of steel at electroslag welding [7]. In some cases, however, heat treatment can induce also unfavourable effects. At welding of refined steels much trouble is often caused by "softening" of the zone heat to the temperature range A_1-A_3 through the effect of troostite precipitation - see Fig. 4.

Structural changes in the welded joint zone connected with polymorphism

Welding of structural steels is, before all, determined by the $A \rightarrow F$ and particular $F \rightarrow L$ transformation during welding. As, in addition to maximum affecting temperature, the manner of austenite disintegration is affected also by cooling rate, for individual steels the diagrams of continuous austenite disintegration under conditions of welding are plotted. Therefore, instead of cooling rate of disintegration time interval more often the transformation temperatures are plotted depending on the cooling time between the temperatures

of 800 to 500 °C, or cooling rate at the temperature of 300 °C. These disintegration diagrams differ from classical ones as at disintegration the conditions of austenite homogenization /temperature gradient in HAZ/ are not given, and disintegration takes place in the zone with unequal primary grain size and under the existence of stress gradients. These gradients not only stabilize the residual austenite, but increase also the carbon concentration in it /3/. One of the methods enabling to determine weldability of steel is the plotting of austenite disintegration diagram "in situ" in the welded joint zone. Another method of imitating the thermal-strain welding cycle under precisely defined conditions is e.g. on the Japanese apparatus Thermorestor-W /Fuji Ind. Electronic/. In Fig. 5 a record of the so-called strainless cycle characteristic for the over-heated part of HAZ is shown, indicating in addition to the temperature curves also the transformation points. Final mechanical properties of specimens with imitated cycle helps us to choose such welding parameters which are optimum for welding of a given steel.

Structural changes, particularly those of high-strength steels, are most often the main cause of crack formation in welded joints. The analyses results show that 90 per cent of cracks in welded joints are cold cracks. Apart of the processes leading to formation of martensite, the susceptibility of steel to cold cracks formation is given by the amount of diffusion hydrogen in weld metal and by the solidification intensity K of a weldment.

The crack occurrence can be prevented not only by the change of welding parameters and decrease of diffusion hydrogen content, but also by the choice of correct pre-heating and post-heating temperatures, or welded joint annealing respectively.

In experiments evaluating the susceptibility of steels to cold or hot cracks formation the cracking tests are used. Some of these tests are plotted in correlation diagrams.

E.g. Ito and Bessyo /8/ give the so-called cracking parameter

$$P_c = C + \frac{Si}{30} + \frac{Mn}{20} + \frac{Cu}{20} + \frac{Ni}{60} + \frac{Cr}{20} + \frac{Mo}{15} + \frac{V}{10} + 5B + \frac{H}{60} + \frac{K}{40} \cdot 10^3 \text{ \%}$$

where C, Si, Mn are weight percentages of individual elements in steel,

H is the amount of diffusion hydrogen in ml/100g of weld metal,

$K = \frac{e}{600}$ is solidification intensity factor /kgf/mm.mm/, which for butt joints can be expressed by help of plate thickness e /mm/.

The necessary pre-heating temperature is then calculated from the relation:

$$T / C / = 1.440 \times P_c - 392$$

In more detail, this method of determining the pre-heating temperature is discussed in /9/.

Precipitation processes and their effect on weldability of steel. At welding of mild, micro-alloyed or low-alloyed steel precipitation is significantly affected.

Two stages ^{of} precipitation processes can be discussed, .viz.

during heating and during cooling down.

Though the changes in distribution and density of dislocations through deformation welding cycle will be discussed later, it must be mentioned here that the level of precipitation in the welded joint zone is affected not only by thermal but also by deformation welding cycle.

In zone heat affected below the temperature of A_1 marks of aging are most often encountered. As shown by statistic survey of WUZ-2S weldability tests on mild rimmed or semi-killed steels, aging can be a limiting factor at welding of these steels. Aging is before all a process of redistribution of interstitial atoms of nitrogen and carbon /10/ at which precipitation of ϵ'' nitride or ϵ carbonide occurs.

In welded joint zone the conditions for aging in zone heat affected by the temperatures ranging from 300 to 400 °C, which in over-saturated solid solution can induce precipitation of ϵ -phase are given. This effect is observable during heating of HAZ. At the temperatures below A_1 , on the contrary, due to the effects of rapid cooling in cooling period of welding heat curve, over-saturated solid solution /C+N/ can form in δ Fe. Together with welding strain cycle this phenomenon can induce the symptoms of strain aging. Such aging is more pronounced when multi-layer welds are used, the effects of individual thermal-strain cycles being here superposed

Though today much knowledge on aging is piled up and we know already how to produce mild steel resistant to aging, anyway from time to time some weldability problems connected

with this process occur in industry.

The symptoms of aging in weld metal should be considered in another way. In some cases /e.g. the "IN" process with increased Al content or vanadium-nitrogen steels/ nitrogen content in steel is deliberately increased up to the limit of its maximum solubility in Fe /0.1%/. Through the effect of specific rolling conditions or heat treatment all nitrogen in steel is bound as aluminium nitride or vanadium carbonitride, this giving the steel good strength and plastic properties. In certain part of HAZ heated above the temperature of 1200 to 1300 °C these precipitates, however, begin to solve and the loosened interstitial nitrogen can cause aging of relevant part of a welded joint. Still more unfavourable consequence of this process impose on weld metal properties. Weld metal is more or less formed by mixed in base metal /10 to 60 %/. This permits nitrogen to enter into weld metal causing pore formation and even aging.

In superheated part of HAZ practically all precipitates contained in steel are dissolved. The only problem which can arise as a consequence of this effect at welding of micro-alloyed steels is over-saturation of primary grains boundaries with micro-alloying elements /Ti, Nb, V/ and precipitation of MX thin films on these boundaries during cooling down. Much more dangerous is the formation of sulpho-nitrides of alloying elements, but this problem was already discussed above.

If, however, higher amounts /0.1 to 0.15%/ of micro-alloying elements /e.g. Ti, V, Nb, Mo/ are used, the Ti Nb, V loosened from the dissolved carbides can increase the hardenability of steel, i.e. prefer the harder disintegration phases of acicular morphologies and increase the hardness values in under-bead zone.

Micro-alloying elements can cause some problems also in weld metal which - as was already said above - is always formed also by mixed-in base metal. While such micro-alloying elements as Ti, Zr, Al can either partially or completely burn in weld metal, others /Nb, V/ remain in weld metal with all negative effects.

The first negative effect can be that on homogeneity of weld metal /formation of hot cracks or micro-cracks/. At welding of micro-alloy steels on the surfaces of opened hot cracks or micro-cracks the presence of thin TiX or NbX films is observed. The second negative effect is precipitation of very disperse NbC particles on dislocations which increases the steel's brittleness /weld metal transition temperature/. Therefore, the Nb content in steels intended for welding should be limited to max. 0.04%. In other case, when the unfavourable effect of Nb in weld metal is to be eliminated, alloying filler material /molybdenum or niobium/ should be used.

Precipitation processes themselves /i.e. homogeneous precipitation are observed in welded joint zone very seldom. Such a case is observable with multi-layer

welds at welding of vanadium steels or as an effect of segregation at welding of low-alloyed steels. In the first case precipitation of very disperse VX particles in under-bead zone or in some parts of weld metal is concerned which increases heterogeneity of welded joint zone /particularly hardness peaks which often cannot be eliminated even by stress relief heat treatment/. The second case can lead to crack formation in weld metal. In Fig. 6 an electroslag weld hot crack surface /CrMoV steel/ is shown, metallurgical cause of crack formation being here molybdenum segregation and precipitation of Mo_2C .

Though the precipitation processes do not take place during the welding process itself, they cannot be separated from the mechanism of annealing cracks formation. In under-bead zone or in weld metal at welding of CrMoV steels of higher thicknesses favourable conditions for formation oversaturated solutions are given. Precipitates are dissolved. During stress relief annealing precipitation of disperse VC in ferritic grains takes place simultaneously. This precipitation together with molybdenum substituted solid solution causes that stress relief goes through CC' creeping on grain boundaries. When deformational ability of grain boundaries is exhausted, cavities formation takes place and under the existence of tensile stresses these cavities propagate to a network of inter-crystalline cracks - see e.g. Fig. 7 of flux weld metal.

The analysis results of annealing cracks formation at welding of CrMoV steels permitted Ito and Nakanishi /12/ to determine a formula for the P_{SR} parameter

$$P_{SR} = \%Cr + \%Cu + 2\%Mo + 10\%V + 7\%Nb + 5\%Ti - 2$$

which characterizes the susceptibility of steel to formation of annealing cracks. If $P_{SR} \geq 0$, steel is susceptible to formation of such cracks.

Evaluation of the effect of deformation welding cycle

It is generally known that the width of zone affected by deformation welding cycle is several times larger than the width of heat affected zone, while at multi-layer welds individual deformation cycles can unfavourably interfere.

Deformation welding cycle causes that aging in the zone below A_1 has a character of deformation aging. We are of the opinion that the deformation welding cycle affects mostly the changes in HAZ below A_1 or in the temperature range A_1 - A_3 respectively, but it does not affect structural properties in the zone heated above the temperature of A_3 . Deformation welding cycle may, however, be the main cause of inhomogeneities ranging from hot cracks and micro fissures through lamellar tearing up to cold cracking. Unfavourable configuration of dislocations, increase of their density, and particularly polygonization processes can affect negatively the plasticity of welded joint zone.

Conclusion

When weldability of steels is evaluated from the view - point of physico-metallurgical changes, then it can be stated that the whole set of welded joint properties and behaviour of steel during welding can be interpreted from the changes taking place in structure and substructure of the steel. Knowledge of these changes have therefore great importance at feed back, i.e. such intervention to steel production, its chemical composition or welding parameters which would eliminate their unfavourable effect.

Captions

- Fig. 1. Transition zone at welding of mild steel. Sulphides do not dissolve up to the solidus temperature. 250 X.
- Fig. 2. Transition zone with dissolved sulphides and formation of hot micro-figures . 250X.
- Fig. 3. Niobium sulpho-nitrides in Nb-micro-alloyed steel after over-heating to 1350°C:
a/ extraction carbon replica,
b/ scanning microscope.
- Fig. 4. Troostite precipitation in A_1-A_3 heat affected zone at welding of refined steel. 250X.
- Fig. 5. Temperature, deformation and stress records /readings/ at simulation of welding thermal cycle on Thermorestor - W apparatus.
- Fig. 6. Hot crack surface with Mo_2C precipitates. Extraction carbon replica.
- Fig. 7. Annealing cracks in submerged-arc weld metal in CrMoV steel. 250X.

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Fig. 1



Fig. 2



Fig. 3a



Fig. 3b

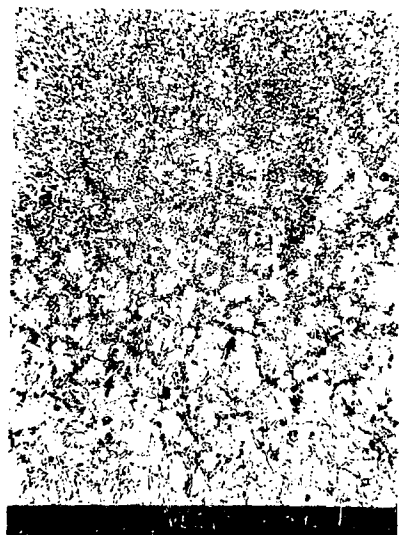


Fig. 4

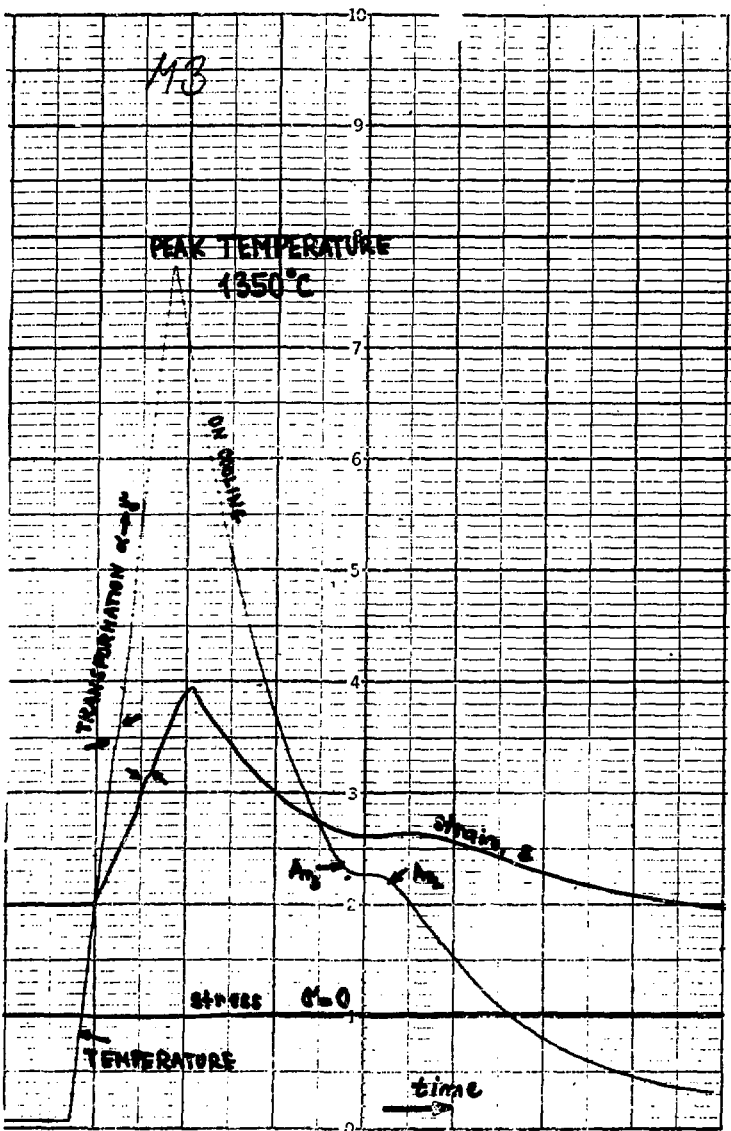


Fig. 5

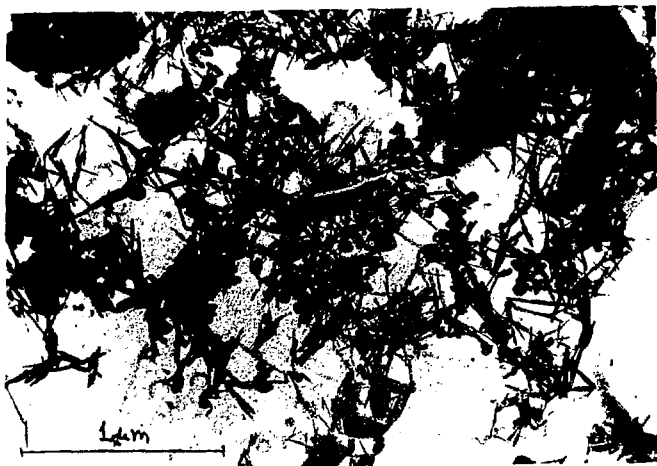


Fig. 6

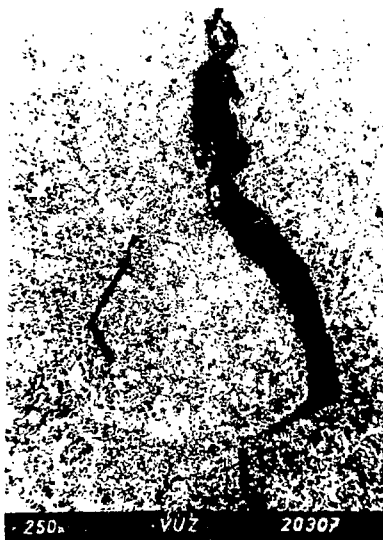


Fig. 7

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FILLER ALLOYS FOR BRAZING IN POWER INDUSTRIES

In current as well as nuclear power industries, in case of fabricating components or equipments, in many cases, due to operative or metallurgical reasons the brazing process is to replace welding. On the relevant brazing alloys high demands are made. Not only the joints are to be of good carrying capacity at elevated service temperature, they must moreover prove to sustain oxidizing and corrosive environment. Such requirements are not met with current brass or silver filler alloys. Suitable are the so called "high temperature brazing solders" on base of nickel. The brazed joint strength is besides the solder and/or parent metal quality, the kind and conditions of the brazing process itself, influenced also by type and dimensions of the joint /1/.

The present contribution will point at the separate qualities and characteristic features of nickel based solders which presently are the mostly used. Also some brief remarks concerning the relevant brazing method will be given.

Solder qualities and properties

The development of nickel base solders has begun some 15 - 20 years ago in the USA and Great Britain first in the aircraft industry /2/. Recently their ap-

plication has propagated over further industrial countries /3, 4/.

Tab. 1 tabulates characteristic features as well as designation of solders to Microbraze Wall Colmonoy and Dewrance Metals companies.

The characteristic feature of the nickel base solders is their powder form of 0.01 to 0.03 prepared sizes. They are however used in form of paste mixed with organic base bonding agent coated on cardboard by means of glue. These solders are substantially less expensive than silver solders.

In Fig. 1 some types of solders are compared from the point of their brazing properties, mechanical strength and resistance to oxidation.

The boron alloyed solders are not suitable for nuclear reactors as boron shows a considerable absorption capability for neutrons as well as elevated diffusion and solubility towards parent metal. Solders of this category, however, are of high strength and are suitable namely in case of heavy sections. The solder micrograph depends on the gap width. With gaps lesser than 0.06 mm the solder structure consists of solid solution - Fig. 2, - connected with improved mechanical properties namely at elevated temperatures - up to 10^4 hrs. exposition /6/. Brazed joints show even a good oxidation resistance /5/ and good fatigue properties /6/.

To be able to apply these solders for joining gaps of more than 0.05 mm, alloys of lower B and Si content, eventually with W or Co addition are to be chosen.

Solders with addition of phosphorus dispose of the lowest brazing temperature. Due to their low diffusion capability and/or mutual solubility with parent metal they are suitable above all for joints in thin sheets.

With solders with Mn addition no high oxidation resistance, nor elevated creep strength in the joint can be expected. Also these solders are suitable for thin sheets. This filler metal can be fabricated in form of wire or foil.

The newest types of solders are based on Mn - Cu - Ni - Co. They are suitable for brazing wider gaps /up to 1,5 mm/. The joints show good toughness, however their applicability in case of elevated temperatures is not so good as with nickel base solders. These are applied for service temperatures up to 800°C. In the present time solders are being developed for higher service temperature - approx. 1100°C.

Brazing processes

With respect to the relatively high melting point of the nickel base solders it is mostly suitable to braze in furnace with hydrogen reduction atmosphere of -40°C to -75°C dew point, or in 10^{-3} Pa to 10^{-4} Pa vacuum, according to parent metal. With boron alloyed solders, cracked ammonia cannot be used as it combines to N_2B_3

preventing the brazing process. Mn alloyed solders cannot be vacuum brazed. The joined materials containing more than 0.5 % Ti, or Al when brazed in reducing atmosphere shall be provided on the brazed faces with a 5 to 30 μ m nickel layer as the arised surfacial oxides cannot be reduced by hydrogen.

Conclusion

The nickel base solders are of good brazing properties and are cheaper than the silver ones. Their merit lays in the fact that their remelting temperature is 150°C to 200°C higher than the original brazing temperature. For that reason these solders are used in power industry for brazing various components and products as heat exchangers, pressure vessels, pipings etc. /7, 8/ made in heat-resisting and creep-resisting steels, stainless steel, Ni, Mo, W, Ta and their alloys.

Captions

Fig. 1. Brazing as well as chemical properties of solders and mechanical features of brazed joints made in 18/8 stainless steel, in dependance on the solder quality.

a/ - flushing of the solder /mm²/

b/ - capilarity of the solder /mm/

c/ - shear strength of lap joint at - 20°C

/thickness of base metal 2 mm, 6 mm lap/

/MPa/

- d/ - like c/ at +600°C
- e/ - shear strength of a step joint at -20°C
- f/ - oxidation of solder at -600°C after 2000 hrs /+mg/.

Fig. 2. Microstructure of a brazed joint in 18/8 austenitic steel - VÚZ NiCr10FeSiB-solder

a/ - with 0.04 mm gap the braze metal consists of solid solution of 231 HVM hardness.
Magn. 300x

b/ - with 0.09 mm gap in the mid part of the joint eutecticum is present showing 651 to 1000 HVM hardness and elevated Cr and Si content. Magn. 300x.

The braze metal is bordered by narrow bands of separated carbides contacted by precipitates along grain boundaries.

Table 1. Characteristic data of nickel - base solders

- 1 - Basic chemical composition /%/
- 2 - working temperature /°C/
- 3 - brazing atmosphere
- 4 - resistance to oxidation /°C/
- 5 - brazing atmosphere: A - dry hydrogen,
B - vacuum, C - cracked ammonium, D -
- exoterm. atmosphere

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Table 1. Characteristic data of nickle-base solders

Ni	Basic chemical composition (%)							Solidus temperature, (°C)	Liquidus temperature, (°C)	Working temperature, (°C)	Resistance to oxid. atmosp.	Microbr-raz NB	Dewrance Metal CM
	Cr	Si	B	Fe	G	W	Co Mn P						
73	14	4,5	3,0	4,5	0,7	-	-	1040/970	1120	A, B	1200	125	56 HC
75	15	4,5	3,0	3,0	0,06	-	-	1070/1010	1120	A, B	1100	10	56 LC
82,5	6,5	4,5	3,0	3,0	0,15	-	-	936/970	1035	A, B	980	LM	53
82,5	10	3,5	2,0	3,5	0,45	-	-	1160/970	1190	A, B	920	160	-
87	5	3	1,0	1,0	0,2	-	-	1180/970	1200	A, B	760	180	-
83	11,5	3,5	2,5	3,5	0,5	16	-	1195/970	1175	A, B	1200	170	-
88	10,0	3,5	2,5	3,5	0,4	12	-	1095/970	1170	A, B	1200	171	-
77	7	4,5	3,2	3	0,1	6	-	1040/975	1120	A, B	980	200	-
71	21	3,0	3,0	-	0,7	10	4,5	1120/1040	1200	A, B	1200	300	-
85	15	-	3,5	-	0,1	-	-	1055/1055	1175	A, B	980	150	-
90	-	4,5	3,5	-	0,06	-	-	1030/985	1040	A, B	980	130	52
94	-	3,5	2,0	-	0,06	-	-	1060/990	1120	A, B	800	135	50
71	19	10	-	-	0,06	-	-	1135/1080	1190	A, B, C	1100	30	60
57	3,5	2,5	1,0	1,0	0,1	-	35	1065/980	1085	A, B	810	230	52 SP
89	-	-	-	-	0,06	-	11	875/875	980	A, B, C, D	760	10	Ni/P
77	13	-	-	-	0,06	-	10	890/890	980	A, B, C	850	50	-
70	-	7	-	-	-	5Cu	23	1010/980	1065	A, B, C	800	65	-
50	4	-	1	-	0,1	-	45	1060/995	1095	A, B	650	220	-
87	4	4	2,5	1,5	0,1	-	-	1090/1060	1170	A, B	800	V0Z	Blava
51	10	4	2,5	2,5	0,4	-	-	1040/990	1100	A, B	850	V0Z	Blava

breazing atmosphere: A- dry hydrogen, B- vacuum, C- cracked ammonium, D- exotherm. atmosphere

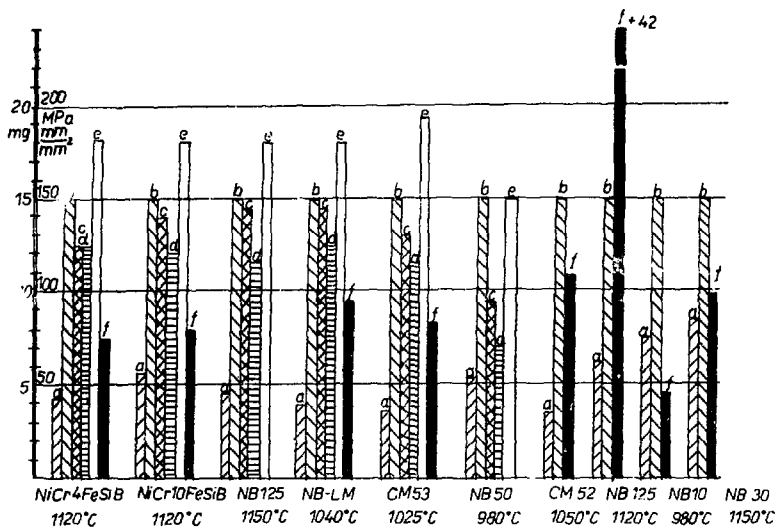


Fig. 1

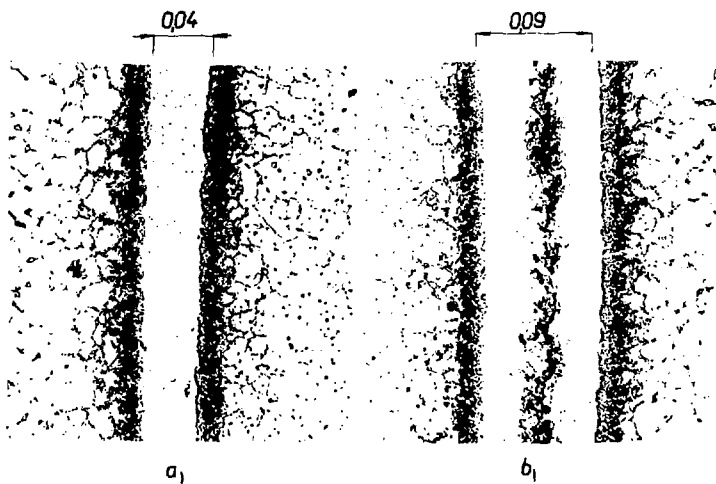


Fig. 2

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ELECTRODES FOR MANUAL WELDING IN POWER ENGINEERING

The strained energy balance in the world leads to a permanent endeavour after the improvement of service efficiency and economy of power engineering equipment. One of the ways in which higher efficiency of steam power stations is ensured, is the increase of parameters /temperature and pressure/ of the overheated steam on which this equipment runs. It can be said that with power engineering equipment made of carbon steel the possibility of further increasing of parameters and thus of improving the efficiency are already exhausted. Similar situation is also with low-alloy creep-resistant steels where creepresistance is attained by using such elements which increase the strength of ferrite or, possibly, formcarbides which block the slip lanes. Low-alloy steels - with various grades of alloying - cover the exploitation range between 450 to 580°C.

Further possibilities of increasing the parameters above 580°C are offered by high-alloy steels, in which the increased creep resistance is attained through solid solution hardening either by substitution or interstition. The effect of some elements on yield point increase through solid solution hardening of austenitic steels is shown in Fig. 1 /1/.

The mentioned groups of steel cannot be used for fabrication of power engineering equipment unless a suitable welding method is devised for their joining. From

among the existing welding technologies - due to very good technological properties and wide assortment of suitable electrodes - manual arc welding with coated electrodes is thought to be the most suitable.

When proposing chemical compositions of weld metals, a whole complex of problems occurring at welding must be considered. From the viewpoint of the claims for loading capacity, short - and long-term properties of welded joint, it would be ideal if the joint were a homogeneous part of the construction with the same properties in all its cross sections as are those of base metal. Such claims, however, cannot be met and it is a permanent endeavour of welders to approach this ideal as close as possible. A decisive factor, which makes the difference between base and weld metal still more pronounced, is the fact that weld metal is of cast structure which is never subjected to further mechanical treatment and only seldom to heat treatment.

With electrodes for welding of low-alloy creep-resistant steels, one of the decisive requirements is that chemical composition of weld metal is as close to chemical composition of base metal as possible. This severe requirement is affected by diffusion processes at service temperatures as a result of weld/base metal dissimilarity. Specifically, the migration of carbon on the fusion line causes the decrease of long-term strength and plasticity of welded joints. This phenomenon mani-

ffects already at small differences in alloying of weld metals by carbide-forming elements. A further important requirement is a proper balance between the alloying elements and the carbon content. For example, according to Sobotka /2/ the structural stability of CrMoV weld metals was found acceptable only at Mo content up to 0.5 %. At higher Mo contents the decrease in creep resistance of weld metal was found, as a result of carbide reactions acceleration which led to precipitation of M_6C carbide. The presence of coarse particles of this carbide both weakens the grain boundaries and de-stabilizes the dispersion phase of vanadium carbide, this manifesting in weld metal dispersion hardening decrease.

Main difference between chemical composition of base metal and weld metal of corresponding quality consists in lower carbon content in weld metal as compared to C content in base metal. Carbon in consumables always brings about insurmountable difficulties. In weld metals, it increases the content of brittle martensite and thus decreases the plastic properties and increases the susceptibility to brittle failure.

When carbon content in weld metal of low-alloy electrodes is decreased, while the other alloying elements are constant and the same as in base metal, comparable creep resistance of weld metal is not obtained. Lower values of weld metal creep resistance are said to be due to precipitation hardening decrease.

In catalogues of world producers of coated electrodes for welding of low-alloy creep-resistant steels we meet with electrodes chemical compositions of which correspond to commercially available steels except for lower carbon content. Generalizing, these electrodes - depending on the supposed service conditions - are Mo, MoCr, MoCrV alloyed or - in some cases - the effect of Mo is partly substituted by W.

In service, we meet with electrodes giving weld metals of the following chemical compositions:

- type I - 0.5 % Mo
- type II - 0.5 % Cr, 0.5 % Mo
- type III - 0.5 % Cr, 0.5 % Mo, 0.25 % V
- type IV - 1 % Cr, 0.5 % Mo
- type V - 1.75 % Cr, 1 % Mo
- type VI - 2.25 % Cr, 1 % Mo
- type VII - 5 % Cr, 0.5 % Mo

As far as electrode coatings are concerned, in majority of cases the low-alloy electrodes have basic coatings. This type of coating ensures the increased deoxidation of weld metal, while the content of non-metallic inclusions and of hydrogen is low. Fig. 2 /3/ shows the allotment of consumables production for individual methods of welding low alloy creep-resistant steels in Czechoslovakia. Small portion of wires for automatic welding indicates serious problems with mechanization of welding work in this field. The Figure shows also a marked portion of

basic electrodes, while acid electrodes represent only a very small volume.

The difficulties encountered at welding of boiler and over-heater tubes with basic electrodes, directed our research work into the field of rutile electrodes application. At welding of tubes on site, at the beginning and at the end of weld - when the electrode is in a tangential position to base metal - weld pool is not sufficiently protected against the outer atmosphere what with basic electrodes may result in formation of pores. The danger of pore-formation is especially high at depositing of root beads; therefore, a combination of electrodes had to be used - acid electrodes for root layer, basic electrodes for the capping bead. Acid electrodes, however, give low micro-purity and consequently low plastic properties of weld metals. Furthermore, this being also a big disadvantage, for making a joint two types of electrodes must be used and, consequently, two welding technologies and two welding current polarities. Rutile-coated electrodes eliminated all these disadvantages to a rather high extent. The decrease susceptibility of rutile electrodes to pore formation in weld metals is usually explained by different physical properties of slag, and particularly by lower inter-phase stress on the slag/metal contact area, this minimizing the possibility of molten metal contact with the ambient atmosphere. Rutile electrodes, as compared with acid

ones, give better micro-purity of weld metal, especially as concerns the content of oxidic inclusions. A further advantage of rutile electrodes is that they give slag of higher viscosity and higher solidification rate, these electrodes are therefore advantageous for position welding. It must be said, however, that from the viewpoint of purity and plastic properties of weld metal, basic electrodes are the best.

Application of austenitic steel for fabrication of power engineering equipment began at about 1950. The development series of austenitic steels was based on conventional stainless steel of 18/8 /18Cr8Ni/ type. A further stage in development of high-alloy creep-resistant steels included also stabilized steels. The reason for it was that the conventional stainless CrNi based steel, when exposed to temperatures above 450°C, was very unstable, especially in connection with precipitation of $M_{23}C_6$ chromium carbide on the grain boundaries. When using stabilized steels, however, cracks are often found on transition zones of welded joints; therefore, in recent years, a growing effort to use un-stabilized steels is observed, particularly the steels of 08Cr17Ni-12Mo2 or 08Cr18Ni10 type. The main advantage of the first type is the increased creep resistance, while the second one is more attractive because of lower price, though a partial decrease in creep resistance must be counted with /4/.

The proposals for chemical compositions of consumables for welding of high-alloy CrNi austenitic steels are affected by metallurgical weldability of steels which is conditioned by the following phenomena:

1. Precipitation of chromium carbides results in susceptibility of weld metal or heat affected zone to formation of intercrystalline corrosion.

2. Formation of the intermediate phase δ /under certain conditions of structural composition and within a certain temperature range/ resulting in embrittlement of steel or weld metal.

3. The susceptibility of weld metal to hot cracking.

The problem of intercrystalline corrosion of δ -phase formation has been so profoundly investigated that nowadays these phenomena can be completely eliminated if at the choice of steel, consumables and welding technology the theoretical knowledge is correctly applied.

From the viewpoint of welding the susceptibility of austenitic weld metals to hot cracking may be considered as the most serious problem which with pure austenitic weld metal has not been successfully solved yet. Hot cracks in weld metals are very dangerous also due to the fact that they are of microscopic character and can be revealed neither by current visual tests nor by laboratory ~~non~~ destructive tests.

According to contemporary theories /5, 6, 7/, the susceptibility of weld metals to hot cracking is defi-

ned by insufficient deformation ability of weld in the brittleness temperature interval.

The brittleness temperature interval is affected by the tensile strength and the width of interval within which the decrease in tensile strength is observed. In pure austenitic CrNi steels, and weld metals in particular, specific phenomena of grain boundaries create suitable conditions for hot cracking. The tensile strength decrease in the critical temperature interval and extension of solidification interval are affected by the presence of low-melt intercrystalline, often continuous, films formed from segregated components of oxides, silicates, sulphides, phosphides, etc. /8/. These segregates, together with elements with limited solubility in austenite, form under suitable conditions the eutectic mixtures which represent weak points in intergrain cohesion and influence thus the hot cracking phenomenon.

At present, it is generally known that if chemical compositions of electrodes' weld metals are counterbalanced by the ratio of austenite-forming and ferrite-forming elements in such a way that weld metal, in addition to austenite, contains also certain amount of ferrite, the susceptibility of weld metal to hot cracking is substantially decreased.

For creep strength properties of weld metals it would be most suitable if the structure were of monophase - austenitic - character. In such a case, the prob-

lems with decrease of welded joints plastic properties after long term exposure to service temperatures would be reduced. Anyway, owing to the above mentioned reasons, electrodes with double-phase austenitic-ferritic structure of weld metal should be developed.

In spite of the fact that the resistance of austenitic CrNi weld metals to hot cracking grows up to 30 % ferrite content, it is advisable, if not necessary, to reduce its upper limit in accordance with recommendations of several authors, to 5 %. This restriction follows from the fact that at higher ferrite content than 5 %, marked increase in hardness and decrease in plastic properties are observed in weld metal after its dwell at elevated temperatures - above 550°C. These changes are the result of δ -ferrite transformation to hard and brittle δ -phase.

The lowest ferrite content which still is able to affect relevantly the resistance of weld metal to hot cracking is supposed to be 2 %.

In majority of cases - unless the resistance of weld metals is not ensured in another way /e.g. by claims for very high purity or specific contents and ratios of alloying elements/ -for welding of creep-resistant austenitic steels such electrodes are used which in weld metal ensure ferrite content within the limits from 2 to 5 %.

The required amount of ferritic phase in weld metal can be attained in two ways.

1. By using a standard wire core and variable coating mixture.
2. By using the constant coating mixture and core wire with specified chemical composition and ferrite content.

Each of the above methods has both advantages and disadvantages. In Czechoslovakia, laboratory production of electrodes with constant coating mixture has been already proved to be possible. Application of this method claims not only that core wire has the prescribed chemical composition but also that ferrite content is kept within the limits $\pm 1\%$ of the prescribed value.

As concerns the type of coating mixture used for high-alloy creep-resistant electrodes, in majority of cases - like with low-alloy electrodes - basic coating is used, while rutile-coated electrodes represent only a small fraction.

With basic coatings, one of the constituents - viz. calcium carbonate - must be kept in such a ratio to other constituents that the resulting slag is of alkaline character. Along with calcium carbonate an important constituent in the coating is fluorite $/CaF_2/$ which ensures that the melting point of slag with high CaO content is decreased - Fig. 3 /9/, this helping to good bead forming. Basic electrodes form gaseous atmosphere rich in CO and

CO₂ which protect weld metal from atmospheric oxygen and bind sulphur enables to nitrogen. The ability of basic slag to obtain weld metal of high purity and thus of good quality. Basic coatings are therefore used in all cases where high purity and good mechanical, especially plastic, properties of weld metal are required.

For welding of creep-resistant steels intended for over-heater tubes of 08Cr17Ni12Mo2 material /corresponding to the AISI 316 H steel/, the electrodes with basic coating yielding weld metal on 18-8-2CrNiMo basis have been developed in Czechoslovakia. Lower chromium and nickel content in base metal shifts in the Schaeffler's diagram the dispersion zone of weld metal to a lower zone of austenite - Fig. 4. This concept of chemical composition is characterized by higher structural stability as compared with that of 17/12/2 CrNiMo weld metal.

In conclusion, I would like to say that the existing theoretical works as well as practical experience in the field of weld metals from electrodes for welding of carbon and low-alloy steels have already reached the peak. Further relevant increase of creep resistance of weld metals can in this field hardly be expected. With electrodes for welding of high-alloy creep-resistant steels, however, in the near future substantial changes in compositions as well as the improvement of creep resistance of weld metals may be expected also due to

the fact that core wires for production of these electrodes are nowadays made of electroslag or vacuum remelted ingots.

As concerns the concept of coating mixtures for electrodes for severe conditions, hardly any serious change may be expected. Rutile, but mainly basic, coatings will most probably be used until a more suitable type of binder than is the water glass used nowadays will be discovered.

List of Figures

Fig. 1. The effect of some elements on yield point increase in austenitic CrNi steels.

% váh. = wt. %; obsah legujúceho prvku = alloying element content; intersticiálne spevnenie = interstitial hardening; substitučné spevnenie = substitutional hardening; zmena v tahu = change in tensile strength

Fig. 2. The allotment of consumables produced in Czechoslovakia for individual methods of low-alloy creep-resistant steels welding.

elektrody s bázičným obalom = basic coated electrodes; elektrody s kyslým obalom = acid coated electrodes; drôty pre zváranie kyslíko-acetylénovým plameňom = wires for oxygen-acetylene flame welding; drôty pre automatické zváranie pod tavivom = wires for automatic submerged-arc welding.

Fig. 3. The effect of CaF_2 addition on melt point decrease of CaO based slag.

tav = melt

Fig. 4. Structure zones of some CrNi austenitic base and weld metals intended for over-heater tubes production.

austenit = austenite; martenzit = martensite; ferit = ferrite.

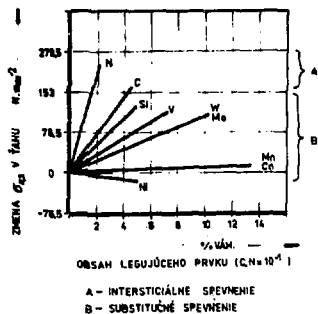
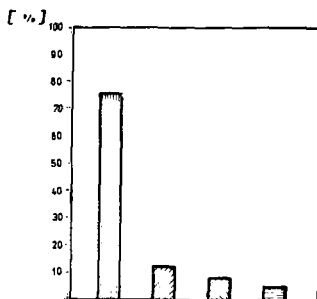


Fig. 1



- ELEKTRODY S ŤAZKÝM ODBALOM
- ELEKTRODY S KYSLÝM ODBALOM
- DRÓTY PRE ZVÁRANIE KYSLIKO-ACETYLÉNOVÝM PLAMENOM
- DRÓTY PRE AUTOMATICKÉ ZVÁRANIE POD TAVIVOM

Fig. 2

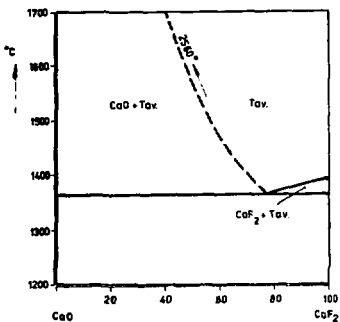


Fig. 3

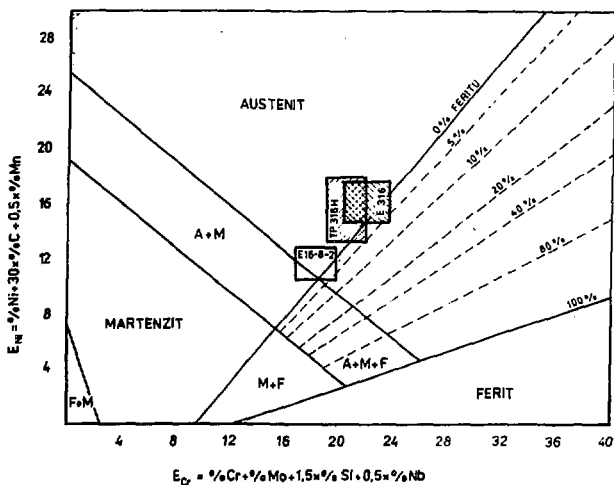


Fig. 4

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FILLER MATERIALS FOR GAS SHIELDED WELDING IN POLAR INDUSTRY

Introduction

The method of electric arc welding of steels in shielding atmosphere of active gases / CO_2 , $\text{Ar}+\text{CO}_2$, $\text{Ar}+\text{O}_2$ / with consumable electrode, referred to as MIG or MAG, has found wide application in industrial practice. The reasons are generally known: high melting rate, overcoming the possibilities of manual welding with coated electrodes, quality of welded joints meeting the requirements of the most challenging applications, but above all the possibility of mechanisation or automation of production operations ensured by welding.

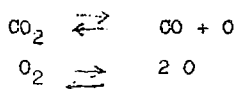
In this lecture, we shall speak about filler materials for this welding method, with respect to the importance of welding wire, which is one of the main factors affecting the final result of welding.

The problems of welding wires for MIG welding

The quality of a weld deposited by this method is affected already by the fact that the welding wire is continuously and mechanically fed to the point of welding. Its clean surface, constant diameter and suitable hardness are important from the viewpoint of weld defects prevention /inhomogenities, cold joints, lack of penetration etc./.

From the viewpoint of the weld properties, the metallurgical effect of the chemical basis of the welding wire is substantial, as the weld is formed in conditions of a very dynamical process of wire melting and metal transfer in the medium of dissociated shielding atmosphere and intense ionisation in electric arc of high temperature.

The basic requirement for chemical composition of welding wires comes out from the necessity to prevent the shielding atmosphere with oxidizing effect from affecting the basic elements /Fe, C/ of the chemical basis of the welding wire in the molten pool or during the transfer of metallic drops into it. This effect of the shielding atmosphere /CO₂/ or of its component /O₂ in mixtures Ar+CO₂, Ar+O₂/ is the consequence of dissociation in the conditions of high temperature of the arc, according to the known equations:



The welding wire must be able to bind this free oxygen by its chemical composition and this way to prevent the formation of oxides. In this function, above all Si and Mn find application. Welding wires are "over-alloyed" by them, to compensate their losses by burn-through. In some cases the wires contain also Al, Ti or Zr as deoxidizers.

Si- and Mn-based wires are used for welding of mild carbon steels. They are, however, suitable also for some applications with the use of low-alloy steels with higher strength. Enrichment of the weld metal by alloying elements from the base material due to mixing enables in these cases to obtain strength values in the weld which correspond to the steel welded. In case of larger thicknesses welded, requiring multilayer welds it is, however, necessary with these steels to use wires with the addition of further alloying elements, as the effect of mixing would not be sufficient here. Such wires are produced either as full, but also in the form of tubes filled with ferro-alloys.

According to their oxidation rate, the alloying elements used can be divided to "constant" and "active" ones. The thermodynamic laws governing oxide formation can be expressed by the free energy ΔF necessary for their formation. Its relative values represent affinity of the elements to oxygen and are, depending on temperature, represented by lines in the diagram shown in Fig. 1 /1/. At given temperature the elements will be the more willing to oxidation, the lower are the lines of their free energy ΔF in this diagram.

The following elements can be considered as "active": /Fe/, /C/, Al, Zr, Si, V, Mn, Nb. The "constant" ones are then: Cr, Mo, P, S, Co, Ni, Cu, /Fe/. This division is confirmed also by the coefficients of burn-through

of elements during the transfer of molten metal from the wire into the molten pool - Tab. 1 /1/. As can be seen, with the "constant" elements the loss by burn-through is less than 10%. Iron can be classified as a "constant" element only if it is shielded by the present deoxidizers /Si, Mn/ from oxidation. Therefore it is considered as an "active" element in the diagram of Fig. 1.

The effect of welding wire chemical composition on the weld metal properties

The use of the specific effects of individual elements from the chemical basis of welding wires must respect their affinity to oxygen from dissociated shielding atmosphere, as was mentioned above. The "constant" elements can find application practically by their whole content. Their effect on the steel properties is generally known and we shall not deal with it concretely. It will be more purposeful to mention some of the "active" elements, with respect to their interactions with the shielding atmosphere components.

Iron. Its oxide - FeO , formed under oxidation conditions, would reduce the present FeO : $FeO + C \rightleftharpoons Fe + CO$, and thus conditions for pore formation in the weld due to the existence of CO would be formed. These reactions are prevented by help of elements with deoxidizing effect with higher affinity to oxygen than that of Fe and C.

Silicon. It is added to welding wires in the amount up to 1%, if they are to be used for CO₂ welding. If the mixtures Ar+CO₂ or Ar+O₂ are used, its amount is reduced to 0.5 to 0.7%. Too high Si content increases the hot cracking susceptibility of weld metal.

Manganese. Itself, it has low deoxidizing effect, but in co-action with silicon it forms easily removable slag. Binding of silicon prevents hot cracking and improves mechanical properties of the weld metal. It is added into the welding wire in the amount of 2.0%, so that there is usually 1 - 1.5% of Mn in the weld metal.

Aluminium. It is a more effective deoxidizer than Si. It is added into wires for welding of steel with high oxygen content, or of rusted plates, in the amount of up to 0.7%, to prevent popre formation /2/.

Vanadium. This is a strongly carbide-forming element, which is used to ensure strength properties of weld metal after annealing. With respect to the possibility of precipitation in the form of carbonitrides with unfavourable effect on notch toughness, its content in the wire should be limited to 0.15% /3/.

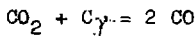
Niobium. It is added to stainless steel wires as carbide stabilizer. It refines the grain of C-Mn steels.

Titanium has a strong deoxidizing effect. Literature refers to its favourable effect on weld metal notch toughness.

In connection with this, its content of 0.15% in the welding wire is considered as optimum, or of 0.03 - 0.04% in the weld metal /4/. It depends, however, also on the content of Al and on that of the alloying elements increasing the quenching ability /Cr, Ni, Mo/.

On the other hand, precipitation in the form of TiX /similarly also Vx and NbX/ would be reflected in the weld metal by increase of the notch toughness transition temperature /3/.

Carbon. When compared to other elements, its action in the weld metal is a certain exception. If the shielding atmosphere is formed by Ar and O₂, carbon oxidates and CO is formed. If CO₂ is the shielding gas, where also CO is present in the atmosphere, the reaction between molten metal and the gaseous phase passes towards the equilibrium state, according to the equation



where C_γ is the concentration of C in austenite.

In the case of low carbon content in the weld metal and high degree of dissociation, the reaction goes leftwards and the result is the growth of carbon content. With higher carbon content, the reaction will go in the opposite direction and the carbon content will decrease, and this way it goes to the equilibrium value 0.07-0.1% with ferritic, and/or 0.1-0.14% with austenitic steels, irrespective of the original content in the welding wire.

The degree of "losses" of the active elements during metal transfer into the weld depends also on the process type applied, namely on the number of droplets - Fig. 2.

Creep resisting properties of weld metals deposited in gas shielding atmospheres

For the production of power industry equipment of plain steels with working temperature below 425 °C a welding wire for CO₂ welding of SiMn type is produced in Czechoslovakia. For equipments working at higher temperatures /425-580 °C/ generally low alloy creep resisting steels are used, alloyed mostly by Mo, Cr+Mo or Cr+Mo+V. Selection of welding wire for these steel types must come out from their chemical composition and must also consider the effect of the above-mentioned welding process effects on the final composition of weld metal. With these applications, the weld metal properties are moreover affected also by heat treatment of the weldment. General composition of welding wires for these steel types produced in Czechoslovakia is given in Tab. 2 and mechanical properties of their weld metals in Tab. 3 /5/.

Creep resisting properties of weld metals deposited from the Mo and CrMo type wires are on the level of the base metal. With CrMoV type wires they are generally lower - they reach only about 70% of the base metal value /5/.

Conclusion

It can be stated that it is the knowledge of the metallurgical processes taking place in the arc and in the molten metal during MIG welding in active gas shielding which is the important pre-supposition of development of sufficiently wide welding wire scale, this enabling the application of this very progressive welding method also in building of power industry equipment.

Captions

Fig. 1. Simplified diagram of free energy necessary for formation of oxides of some elements /1/

θ - temperature /°C/

ΔF - free energy of formation of oxides of some elements

Fig. 2. Weld metal composition in dependence on metal transfer mode during welding in CO_2 atmosphere /1/

n - number of droplets /1/s/

V - % of element in weld metal

a - unapplied sphere of the welding process

b - process with low number of droplets

c - process with high number of droplets

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ných mikropevných ocelí typu Mo, CrMo a CrMoV
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Table 1. Burn-through losses of the individual welding wire elements in CO₂ welding /1/

a	b	C ^x	Al	Zr	Ti	Si	V	Mn	Nb
		50-235	30-40	30-40	40	50-70	60	60-75	60-70
c		Cr	Mo	P	S	Ni	Cu /Fe/ ⁺		
		90-95	95-100	100	100	95	95-100		99

^xThe minimum value applies to ferritic steels with high carbon content in the wire /0.30%/ - it will decrease to 0.15%. The maximum value corresponds to austenitic steel with low carbon content 0.055% - it will increase to 0.13% in the weld metal

⁺in the presence of deoxidizing elements Si and Mn

a - % of transfer = $\frac{\text{element perc. in weld metal}}{\text{element perc. in welding wire}}$

b - "active" elements

c - "constant" elements

Table 2. Chemical composition of Czechoslovak wires for gas shielded MIG welding of low alloy creep resisting steels

a	b	C	Mn	Si	Cr	Mo	V	d
15020		0.07-	1.2-	0.50-		0.40-		
15110	P-Mo	0.12	1.3	0.75	-	0.60	-	
15121	C-312	0.07-	1.0-	0.50-	0.90-	0.40-		CO ₂ or
		0.12	1.3	0.75	1.20	0.60	-	
15128.5		0.07-	0.90-	0.45-	0.50-	0.50-	0.25-	30% Ar+
/15123/	C-321	0.12	1.20	0.70	0.70	0.75	0.45	20% CO ₂

a - steel

b - welding wire designation

c - directive chemical composition of welding wire

d - shielding atmosphere

Table 3. Mechanical properties of weld metals deposited
by MIG welding in active gas shielding

a		d				e
b	c	σ_{Pt} [MPa]	σ_{Kt} [MPa]	δ_5 [%]	K_{Ic} [Jm ^{-3/2}]	
P-Mo	P-Mo	600	340	23	74	680 °C/2hrs/air
P Mn1Cr1Mo- 26C-AC	C-312	550	380	16	45	700 °C/2hrs/air
P Mn1MoCrV- 24C-AC	C-321	550	400	14	45	720 °C/2hrs/air

a - wire designation

b - according to the proposed standard ČSN 5390

c - according to the producer

d - mechanical properties of weld metal at +20 °C
postweld

e - heat treatment

σ_{Pt} - tensile strength

σ_{Kt} - yield point

δ_5 - ductility

K_{Ic} - notch toughness /DVI/

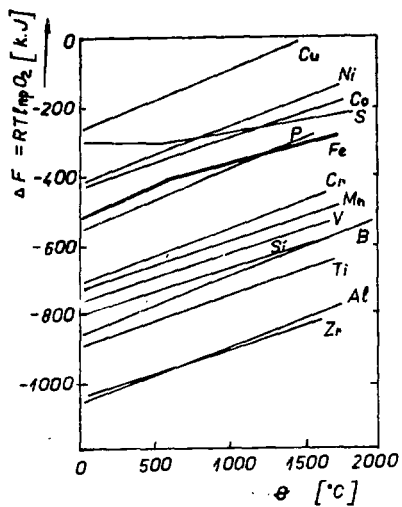


Fig. 1

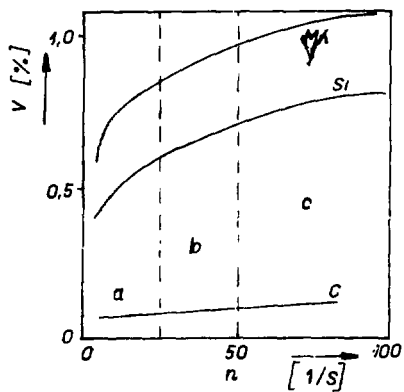


Fig. 2

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POWER ENGINEERING STEELS

A. Introduction

Steels used in power engineering may be classified into several groups. Firstly, the carbon /C_{mn}/ unalloy and micro-alloy steels of which most of the auxiliary equipment is made. Further, the low-alloy Cr_{Mo}, Cr_{Mo}V, MoV or Cr_{Mo}VW steels of which superheater tubes and some other power engineering parts stressed in the creep zone are made. Among the present-day peak steels of this type ranks also the 12%Cr steel with its modifications. Finally, the high-alloy austenitic CrNi steels and Ni alloys which find their application not only in production of steam generators but also in production of power engineering gas turbines.

To power engineering equipment belong also steam and gas turbines with ever increasing outputs. For units of the output higher than 200 MW it is no longer advantageous to use one-piece forged rotors, it is better to make them by welding together several /up to fifteen/ sections. As this new technology imposes certain demands on steel, modified low-alloy rotor steels had to be developed. Of similar character are steels intended for water turbines /Kaplan and Pelton types/. Finally, mainly the nuclear power industry requires special steels for nuclear reactor pressure vessels and for primary circuits.

Some power engineering devices require steels specially designed for low temperature service. Here, some maraging 9%Ni steel have found the field of their application. Basic classification of the above mentioned steels and discussion of the problem of their weldability are the subject of this paper.

B. Theory of steels

Going out from a mono-phase ferritic structure, the $\sigma_{0.2}$ yield point value can be calculated using a modified well-known Hall-Petch relation:

$$\sigma_{0.2} = \sigma_0 + \sigma_S + \sigma_{IN} + \sigma_D + \sigma_d + \sigma_P + \sigma_{Disl}$$

where σ_0 is the so-called friction stress of the lattice α -Fe,

σ_S is the contribution of substitutional hardening α Fe by such elements as Mn, Si...,

σ_{IN} is the contribution of interstitial hardening α -Fe /C, N, O, B/,

σ_D is the contribution of grain boundaries hardening

$$\sigma_D = k D^{-\frac{1}{2}} \quad /D = \text{grain size}/,$$

σ_d is the contribution of sub-grain boundaries hardening

$$\sigma_d = k' d^{-\frac{1}{2}},$$

σ_P is the contribution of precipitation hardening, depending on the size and level of dispersion of precipitated particles,

σ_{Disl} is the contribution of hardening through the increase of dislocation density.

Each particular contribution is influenceable to a rather great extent by different chemical compositions and/or heat treatments of steel.

In fact, there does not exist anything like pure ferritic steels, but only steels with certain portion of eutectoid. Here, it must be mentioned that the higher is the portion of pearlite in the steel, the higher are also its strength properties, the dependence being nearly linear.

With low-alloy high-strength steels the solid solution hardening is ever more often used.

This hardening, applicable both to ferritic and to austenitic steels, goes out from the effect of elements which form neither carbides nor intermetallic precipitates. These elements are not distributed uniformly in the matrix but form "segregates". The segregates are responsible for hardening of the solid solution as they increase the energy necessary for movement of dislocations. The segregation of atoms can take place either through interaction between the dissolved atoms and the imperfections, or between the dissolved atoms themselves. The elastic interaction between the dissolved atoms and the dislocations leads to the Cottrell's atmospheres, while the interaction of the electric field lying between these atoms and the stacking faults leads to the Suzuki's effect. Finally, the interaction between the dissolved atoms themselves leads to a short-range ordering /Fisher/ or to clustering.

Apart from the basic kinds of hardening, which have been already discussed, for the steels of higher tensile strength used in quenched and tempered condition, also the trans-

formation hardening by bainitic or martensitic reactions is used. Decrease of the transformation temperature with the aim to obtain bainite or martensite is the simplest way of increasing the steel's hardness.

Bainite is a constituent of steels formed by decomposition of austenite in the temperature zone between martensite and ferrite and pearlite. This constituent consists of aggregate of ferrite and carbide and partly of stabilized austenite. The morphology of bainite changes progressively with the transformation temperature; the length of particles grows with the decreasing temperature, the resulting structure being thus acicular. Morphologically, three bainite types can be distinguished: massive, upper and lower bainite. The first of them is also called granular bainite /at smaller magnification/ or acicular ferrite /at greater magnification/.

The relation between the structure of low-carbon steels, their yield point and transition temperature is shown in Fig. 1. For the steels of higher tensile strength, quenching and tempering must be used to attain sufficiently low transition temperatures. Fig. 2 shows the relation between grain size and yield point for different microstructures. The main line in this diagram represents the relation between the yield point and grain size according to the Hall-Petch equation.

For the C-Mn-Fe steels the deviation /increase/ of yield point is due to precipitation hardening of MnC. For bainitic and martensitic steels the increase in tensile strength is due not only to dispersive carbide particles

but also to the increase of dislocation density.

With martensitic steels the contribution of interaction between the dissolved atoms of carbon and the dislocations has to be counted with, too.

C. Micro-alloy steels

Through the C_{Mn} steels, like micro-alloy ones are used in power engineering in such cases only when the steel is not stressed in the creep zone /up to about 350 °C/, great majority of the used low-alloy /e.g. reactor or rotor/ steels is produced as micro-alloy Ti or Nb steel, the data given here may thus have a wider validity. Under the term "micro-alloy steels" we understand C_{Mn} steels in which the content of micro-alloying elements /Al, Ti, Nb, V, Zr/ does not overreach 0.1 t.%. In some cases, also higher contents are acceptable /e.g. V up to 0.15% and, on the contrary, in others the maximum content is limited still lower /Nb 0.04%/. Typical ranges of micro-alloying elements used in these steels are:

Al - 0.02 to 0.07% ,

Ti - 0.02 to 0.15%,

Nb - 0.01 to 0.05%,

V - 0.01 to 0.10%,

Zr - up to 0.15%.

The above ranges can vary at some combinations of micro-alloying elements, e.g. NbV, TiNb or TiZr. In great majority of micro-alloy steels the presence of aluminium as a remainder of decaridation process is observed. In our

case, its content is lower than 0.02% /4 to 6 kg, for one ton of melt/, and have not any pronounced hardening effect. On the contrary, in some steels the Al content is deliberately increased up to 0.1% with simultaneous increase of N content up to 0.1% /the "IN" process/. Such steels require that the control of rolling regime or primary heat treatment is very precise to ensure that all interstitial nitrogen is bound to Al in form of nitrides.

The latest development in micro-alloy steels are steels with low carbon content /below 0.1%/, the increased Mn content /up to 1.6%/, and the presence of 0.3% Mo together with max. 0.1% Nb. The MoNb combination shows pronounced increase of strength properties at very good plastic properties values. As already mentioned in the first paper presented at this course, the production of micro-alloy steels requires control of rolling and of cooling regime and, possibly also, of steel heat treatment. The aim of this control is to obtain very fine precipitate with sufficient degree of dispersion. If the process of precipitation realizes in form of thin MX films or sulphonitrides on primary grains boundaries, then the plastic properties of steel show marked decrease.

The bond of micro-alloy to carbon in form of carbide or carbonitride enables - at simultaneous decrease of carbon content - to decrease the portion of pearlite in the steel. In this way, steel with reduced pearlite content or even pearlite-free steels can be manufactured.

It is necessary to say that when micro-alloying is used, the chemical basis of sulphidic inclusions in steel changes, too. A part of inclusions of sulphidic type is bound not to Mn but to the used micro-alloying element. It can also be said that a part of manganese sulphides is further alloyed by microalloying element.

This effect is typical for Ti and Nb, but aluminum sulphides are known too. It is important that these sulphides have different physical properties, especially the decreased melting temperature as compared to MnS. As a consequence, micro-alloy steels are more susceptible to micro-hot tearing or formation of lamellar fissures as compared to conventional C-Mn steels. Consequently, the use of micro-alloying necessitates that the content of impurities in the steel is reduced. Within the IIV it is accepted nowadays that steels are not susceptible to lamellar tearing during welding if the reduction in area value of a specimen in the plate thickness direction is higher than 20%.

Speaking about the weldability of micro-alloy steel, it must be mentioned herethat it is better than that of conventional C-Mn steel of the same strength category. The weldability improvement^{ement} is due particularly to decrease of carbon content below 0.15% as compared to 0.20 to 0.25% in conventional steels. Further, the effect of carbon content decrease is favourable from the viewpoint of carbon equivalent value and cracking parameters, as well as from the viewpoint of resistance welding. With this tech-

nology, all weld metal is formed by remelted base metal, and high carbon content increases the range between the liquidus and solidus of the steel, this resulting in crystallization cracks or cavities in weld metal. This enables to weld micro-alloy steel of 600 N mm⁻² strength category, i.e. with carbon equivalent value up to $\approx 0.45\%$ and of 25 mm in thickness without pre-heating or, in case of fillet joints or severely stressed parts, with pre-heating to 80 - 120°C.

At welding of micro-alloy steels, particularly those with reduced pearlite content, comparatively narrower heat affected zone is observed. In the over-heated part of the HAZ, formation of acicular decomposition structure is observed instead of rough ferrite-pearlitic structure with λ -texture which - from the viewpoint of fatigue and plasticity properties - is far less advantageous.

Therefore, micro-alloy steels can also be used in dynamically loaded welded structures, where conventional C Mn steels are not recommended.

Consequently it can be said that micro-alloy steels are more advantageous than conventional C Mn steel of the same strength class. Apart from higher susceptibility of micro-alloy steels to lamellar and hot tearing there are also some other factors which necessitate precise control of the welding technology.

Firstly, it is the fact that while in base metal the presence of micro-alloy is favourable, in weld metal /particularly in Nb steel/ carbide or carbonitride bonds to dislocations, thus decreasing the notch toughness value of the weld. It is not a simple problem, and we still do

not know how the mechanism of damaging the impact properties works. With majority of arc technologies a greater part of micro alloy burns out in weld metal; at submerged arc welding, mainly with higher specific heat inputs, the mixing of base metal into weld metal is so great that the present niobium increases markedly the weld metal's transition temperature. This is a reason why with submerged-arc welding technology the heat input must be limited to 25-28 kJ/cm, and when low transition temperature is required, the 0.5 Mo or 1Ni0.5Mo welding wire must be used.

According to the IIW method /documents IX-902 -74 and IX-903-74/as the type test of the HAZ embrittlement by which the weldability of steel is being evaluated, the test of under-bead notch toughness at various welding parameters has been chosen.

When relating the transition temperatures in the HAZ of micro-alloy steels to the transition temperatures of base metal, the micro-alloy steels' weldability evaluation is not very favourable. For instance, a common micro-alloy steel after classical rolling has the transition temperature of T_1 , and after controlled rolling the temperature of T_2 , while $T_2 \ll T_1$. In the underbead zone both conditions of this steel have the same transition temperature T . Now we see that $T - T_2 > T - T_1$, i.e. relative shift of the underbead zone transition temperature is higher after controlled rolling, i.e. the weldability is worse. Such an interpretation is not logical for a given steel type.

D. Steel for use in the creep zone

With majority of steels used in power engineering it is required that their mechanical properties are good also at increased temperatures, i.e. in the creep stress zone. With classical power engineering equipment, such as steam generators or steamlines, the service temperature of the pipe's wall is up to 600°C . In this temperature zone it is important that:

- steel has sufficient creep strength,
- and sufficient resistance to surface oxidation.

The knowledge of mechanism and kinetics of creep enables that ever better steels are produced. Creep at high temperature is affected, apart from steel's basic chemical composition, also by its structure and sub-structure. In addition to precipitation processes - its result of the first importance being the degree of precipitates' dispersion - also the condition of dislocation sub-structure of metal is important.

Creep is characterized by movement of dislocations and diffusion of vacancies.

From the point of view of creep strength, steels can be classified into several categories. /Mention please, that the measurable temperature zone at which creep begins is always above 200°C ./

- Unalloy CrMn steels which from the viewpoint of creep strength can be used up to the temperature of max. 400°C, while their resistance to oxidation is sufficient up to the temperature of 520°C.
- Low-alloy CrMo, CrMoV, CrV, CrMoVW steels, the temperature zone of their application being limited by 580°C.
- Highalloy, e.g. 12%Cr steel and austenitic steels - wall temperature can rise up to 620°C or even higher.

Some steel types applied in Czechoslovakia are tabulated in Tab.1. The basic type of low-alloy creep resistant steel is CrMo steel containing up to 2 1/4Cr and 1% Mo. Usually, however, Cr content in this steel type ranges from 0.5 to 1% and Mo content from 0.4 to 0.6%. Chromium contributes not only to the increase of creep strength but also of heat resistance of the steel; the present molybdenium both hardens the α -Fe solid solution and stabilizes the cementite.

In CrMo steels the carbidic phase is usually precipitated not only in form of alloyed M_3C cementite, but also in form of Cr_7C_3 plates or - particularly after long-term service in creep zone - in form of Mo_2C molybdenium carbide can also be one of the steel constituents presenting some problems at welding. Firstly, it can be a cause of weld metal embrittlement at arc welding. Owing to higher carbon content, at arc welding with higher heat inputs the Mo_2C can precipitate - as a result of segregation - in irregular clusters .

Another case of embrittlement can occur in the H.Z after heat treatment. If together with annealing of welded joints a totally coherent precipitate forms /usually around the temperature of 530°C /, secondary hardening and embrittlement can occur. It is important that this zone is overcome at heat treatment. All the same, it is not impossible that with unannealed welded joints this precipitation stage is reached during their service life. The service temperature of CrMo steels is limited by the value of 500 to 520°C . At higher temperatures, the α /Fe solid solution lacks already sufficient strength which can be regained e.g. by sufficient amount of dispersion precipitates.

This can be attained by adding vanadium to the steel. These CrMoV steels contain, in addition to 1%Cr and 0.5%Mo also 0.2 to 0.35%V. Vanadium forms with carbon /or nitrogen too/ very stable carbide or carbonitride V_4C_3 /VX/. Vanadium carbide, same like other carbides or carbonitrides M_2X /TiX, NbX/, is distinct for the fact that its size is the function of temperature and practically does not depend on the exposure time.

The zone of precipitation of VX in creep resistant steels can be limited by the temperature range between 650 to 750°C . To attain sufficient creep resistance of the steel, it is annealed in the zone of precipitation of VX.

The size of VX particles ranges between 200 to 500 \AA and, apart from geometrical configurations or irregular disks, they are often fibreform configurations. Apart from conventional vanadium alloyed steels /0.2 to 0.4%V/, vanadium is added - though in far less amounts - also to boiler steels

to obtain the required strength properties of heat treated steel. In this case, the vanadium content does not exceed 0.15% and spontaneous precipitation of VX in the steel is practically unascertainable.

The weldability of boiler CrMo vanadium steels is not worse than that of the previous type. However, to have full advantage of favourable effect of vanadium carbides, these steels have comparatively higher carbon content, this requiring use of pre-heating prior to welding within the temperature range of 200 to 300°C. A very difficult problem encountered at welding of the above steels is the increased hardness in the underbead zone. It is usually accepted that from the viewpoint of strength homogeneity, hardness of the underbead zone should not be higher than 20% more than that of base metal. As the hardness peaks at multi-layer arc welding are due to dissolution of vanadium carbides and their re-precipitation after high oversaturation /high affection temperatures in the underbead zone/, it is often impossible to eliminate these hardness peaks even by post welding heat treatment at the temperatures below A_1 . For welding of these steels all known welding technologies can be used. With all arc technologies pre-heating up to max. 350°C is required. The increase in hardness of the underbead zone can be the cause of difficulties also at electroslag welding of thick plates for steam boilers. These boilers are subjected after welding to size calibration /hot rolling/. Owing to hardening of a portion of the HAZ and to comparatively softer weld metal, the overlap of the HAZ by weld metal can occur during rolling, and this can be a cause of false indications of inhomogeneities.

The majority of welded joints is annealed after welding at the temperatures below A_1 . The annealing is not only to decrease the level of residual stresses but also to eliminate the occurrence of unstable decomposition phases in the welded joint zone and, at the same time, to provoke the precipitation of vanadium carbides. In this connection it should be mentioned that increasing of the pre-heat temperature above the temperature of M_S /of the steel/ is not justifiable and is even harmful. It can result in stabilization of residual austenite not only in weld metal but also in portion of the the HAZ. From the surrounding oversaturated decomposition phases the interstitial carbon can diffuse to this austenite, this further increasing its stability.

Then, when welded joints are annealed at the temperature below A_1 , only the transformation $\delta \rightarrow \gamma$ can be expected, while the originated martensite is sufficiently stable. Not all welding technologies require "a priori" annealing of welded joints. For boiler tubes of smaller diameters /up to about 45 mm/ and wall thicknesses /up to about 5 mm/, welded using the electric resistance method and butt joints, not in all cases the annealing of welded joints is necessary.

One of the top quality ferritic steels - intended for the temperatures of 600°C, i.e. for overhater tubes or boiler bodies or steamlines, but also for cast boxes of turbogenerators - is the 12%Cr steel. This steel with its various modifications is used not only in steam generators or turbo-sets but also in Kaplan's or Palton's water turbines.

In the basic 12%Cr steel type the carbon content is low /0.1%/, though in final products its content can be increased up⁺ to 0.20%. In addition to small amount of Mn /0.6%/, and Si /0.30%/,⁺ the steel can contain 0.5%Ni, 0.1%Mo, 0.55%W and 0.3%V. This basic steel is known as AISI 616 or 422. The temperatures M_s and $M_{50\%F}$ for these steels can be calculated according to the equations:

$$M_s /^{\circ}C/ = 832 - 29/\%Cr/ - 39/\%Ni/ - 5/\%Co/ - 36/\%Mo/ - 0/\%Ti/$$

$$M_{50\%F} /^{\circ}C/ = 1,561 - 47/\%Cr/ - 87/\%Ni/ - 28/\%Co/ - 36/\%Mo/ + 114/\%Ti/$$

Chemical composition of the steel should ensure that the amount of δ - ferrite at the austenitization temperature is as small as possible. The addition of 1% of alloy /excluding C and N where 0.1% is supposed/ the δ - ferrite content will vary in this steel type as follows:

Change of	N	C	Ni	Co	Cu	Mn	W	Si	Mo	Cr	V	Al
δ -ferrite	-20	-18	-10	-6	-3	-1	+8	+8	+11	+15	+19	+38

The 12%Cr steel belongs among maraging steels, the acceptable strength and plastic properties of which are obtained only after normalizing and annealing in the temperature range between 66 to 740°C.

With this heat treatment the structure of matrix is formed by the α -Fe solution with uniformly dispersed carbides. When heating the 12%Cr-0.1%C steel beyond the temperature of A_s , austenite originates. This temperature depends on the manner of steel alloying, and ranges between 600 to 740°C. Use of the steel is limited in the creep zone by the temperature of A_s . The effect of various elements on change of the A_s temperature is the following:

Element /1%/ ⁺	Ni	Mn	Co	Si	Mo	Al	V
Change of A_s /°C/	-30	-25	-5	+20	+25	+30	+50

As evident from the above, the steels of 12%Cr type which are intended for service in the creep zone, will be mainly Mo and V alloyed while the same steel intended e.g. for water turbine rotors will be mainly Ni and Co alloyed.

Even though the calculated A_3 temperature indicates in the equilibrium diagram the temperature of beginning of the transformation, it does not mean that at long-term exposure to the temperatures below A_3 the austenite is not found. Small portion of reversion was observed with these steels still at the temperatures of 500°C. This steel type, same like all high-alloy Cr steels, shows below the temperature of 500°C the disintegration of b.c.c. solid solution into two isomorphic phases, one of them being rich in chromium / / and the second in iron / /.

This disintegration can be expected at long-term exposure of steel to temperatures below 500°C. It seems that disintegration of Fe-Cr solid solution to δ and δ' is not the final stage but only an intermediate one of the final equilibrium product, e.g. δ' - phase.

As far as weldability of 12%Cr steels is concerned it can be stated that no significant problems are encountered. All welding technologies, including electroslag, are applicable. Pre-heat temperatures range between 250 to 350°C, and at repair the welds can also be made without pre-heat with manual austenitic electrode. There can be some problems with notch toughness values in the heat affected zone, particularly in the zone heated to the temperatures $A_1 - A_3$.

To improve the notch toughness value in the HAZ, 2 to 5% of Ni is added to the 12%Cr steel. When welding this steel type it is suitable /if the intricacy of the construction and thickness of the weldment permits it/ to decrease the interpass temperature between welding and annealing to the temperature close to M_s / $\sim 150^\circ\text{C}$ /.

In case of 12%Cr steels with low Ni content, tempering embrittlement may be observed when annealing the welded joints in the temperature range between 600 to 650°C. This is not observed with Ni alloyed steels.

If the welded joints are immediately after welding annealed from the interpass temperature, it may happen that big amount of residual austenite in the HAZ, but also in weld metal, will during annealing transform to martensite. In such a case, double annealing should be used; martensite decomposition to sorbitic components will thus occur at the second annealing only. Owing to different requirements of classification societies the presence of martensite in the HAZ need not necessarily reflect in the notch toughness values; It will however reflect in high hardness and especially in tearing at bend tests of welded joints.

From among austenitic creep resistant steels used in power engineering the widest field of application has found the AISI 316 steel. This steel has, apart from good plastic and strain properties which are very important e.g. at fabrication of tubes with quality surface, also good strength properties. At welding of steel of this type the 16-S-2 consumables, e.g. Thermit Micro 82, are used.

2. Low-alloy steels for water and steam turbine rotors

In case of steam turbines fabrication of rotors by welding from several pieces is economical only when the outputs over 150 to 200 MW are concerned. For root beads arc welding or automatic TIG/MIG or electron beam welding methods are used. Further layers /filling/ of welded joint are most often deposited by submerged-arc welding or, in some cases, by manual electric arc welding. At S-A welding, the narrow-gap technology is used with advantage.

Rotor steels are most often of low-alloy CrNi or NiCrMo type; characteristic chemical compositions are given in Table 2. Owing to the required strength properties the steels have higher carbon content /0.20 to 0.25/ than conventional structural steels. To ensure fine-grain structure and, in particular, to attain the required strength properties in as annealed condition, the steels are alloyed by small quantity of vanadium, and in some cases Ti or Nb microalloyed. As the required strength properties are rather high, the forged rotor parts are refined.

The used alloy basis of the steel requires good deal of care not only at welding but also already at production of the steel. Crystallization right to peritectic reaction emphasizes the segregation effects, and consequently these steels must contain lower amount of impurities.

The use of titanium or of AlTi group may increase, at steel's deoxidation, the susceptibility of the steel to hot or micro-hot tearing, if the technology is not strictly observed.

If we realize what the thermal regime of welding is like, then the biggest problem connected with welded joint's soundness is a possible occurrence of:

- a/ hot tearing,
- b/ reheat cracks

The pre-heat temperature ranges between 300 and 350°C.

These values are valid for welding of joint's root which - owing to the necessity of securing the rotor's axial alignment - is often made in vertical position.

Use of higher pre-heat is not possible from the viewpoint of metallurgy: owing to vertical position the upper part of root layer is heated more intensively, and this may result in incorrect forming of weld. Here again must be emphasized that increasing the pre-heat temperature beyond the M_s temperature of the steel is theoretically unjustified.

For further layers /filling/ - which are deposited already in horizontal position - the interpass temperatures can be lower by 50 to 100°C. The use of high interpass temperature /e.g. the same as the pre-heat temperature/ may result - especially in capping beads - in trapping /presence/ of higher amount of residual austenite.

Steels of this type, both in the HAZ and in weld metal, show usually only two transformations, as evident from Fig.3 where a typical 2 1/4 Cr1Mo steel is presented.

Structure of the welded joint zone is in as welded condition formed by heterogeneous mixture of decomposition phases: martensite, lower bainite and residual austenite.

Among these phase, the islands of massive ferrite can occur too. The amount of residual austenite usually does not overreach in the HAZ 10%, but in weld metal it can reach up to 20 to 30%. From the neighbouring unstable decomposition phases carbon can diffuse to this residual austenite.

Carbonization of residual austenite still improves its stability. Thus, it is possible when higher interpass temperature is used - that austenite transforms at annealing to martensite. It is therefore recommended that when welding is finished, weld metal is kept on the interpass temperature /180 to 250°C/ for several hours. During this time, hydrogen diffusion and disintegration of residual austenite to martensite take place. In majority of cases, welding is succeeded by stress relief annealing, the welded joint temperature not being permitted to drop below 180°C /interoperational temperature/. Then annealing follows while the rate of heating to annealing temperature should not be higher than 20°C per hour at the temperatures up to 300°C. The steels used for turbine rotors have such an alloy base which does not increase their susceptibility to cracking at welded joints annealing. The cracking can occur in two temperature ranges: at 100 to 300°C if the heating rate is high and internal stresses are accompanied by thermal stresses induced by the difference between the surface and core temperature of the weldment, and structural stresses induced by disintegration of oversaturated phases.

The second temperature range is limited by temperature beyond about 500°C, and is observed at exhaustion of creep on the grain boundaries. The grains themselves are hardened either by precipitation of MX or substitutionally, so that the relaxation of internal stresses must take place through creep on the grain boundaries.

If the boundary space is occupied by coarser precipitates, e.g. $M_{23}C_6$ carbides, then at tertiary creep cavities or reheat cracks are observed there.

The weldability of rotor steels is limited by the above phenomena. We therefore prefer the steels with higher level of purity and lower content of carbon and alloying elements. This, however, is in contradiction with the required strength properties as well as with the required notch toughness values.

In case of steels for water turbines, the above requirements are still accompanied by a very significant requirement for cavity resistance. From this reason, nearly exclusively 12%Cr or 12%Cr4%Ni steels /already discussed above/ are used for water turbines nowadays.

F. Steels for nuclear reactors pressure vessels

From the viewpoint of historical development, the nuclear reactor pressure vessel steels can be classified into three groups:

I. Steels of the first generation, characterized by yield point lower than 500 N/mm^2 of which the majority of the existing pressure vessels are made. This group includes ASME A 212 - V, A 302 - B, A 350 LF1 to A 350 LF 3, partly T-1, GOST 22K steels, and Czechoslovak modified 13 030 /AlTi/ steel.

All these steels are characterized by high plasticity, high fracture toughness and low transition temperature. Owing to comparatively low strength properties, walls of pressure vessels are very thick /from 60 to 250 mm/ and heavy.

II. In the second generation, Q_T steels of higher tensile strength and yield point between 500 to 900 N/mm^2 are used. These steels are either modifications of steels of the first generation /using such alloying elements which permit Q_T treatment -e.g. A 350 - LF1 to LF3/, or a result of a complex alloying. These include e.g. steels A 542, A 543, A 508, IX-30, GOST 12 Ch 2 MoA, 25 Ch 2MoA, 15 Ch 2MFA, 25 Ch 2 NiMoA, or Czechoslovak refined steel 15 219. Majority of the present-day reactor vessels made of these bainitic or sorbitic steels with good plastic properties and radiation stability.

III. The third generation is formed by perspective steels, and should be used for production of high-output reactors. These are steels with yield point above $1,000 \text{ N/mm}^2$, Q and T and with high claims for pro-

duction and control. This group includes the steels : ASME A 538 /Maraging/, 12-5-3 /Maraging/, HY 100 and HL 130 /T/, HP 9-4-20 CrMo etc.

Chemical compositions of some characteristic steels of all three generation are given in table 3. This table shows that - except small details - the particular steel types of various producers have the same chemical composition.

Owing to requirements for high purity, low defects occurrence, good plastic properties, and resistance to radiation damage, majority of melts are made in electric furnaces with subsequent vacuumizing. Individual parts of reactor vessels are made as forgings or plates. In both cases it requires the use of heavy ingots /70 to 100 tons/^x, production of which pose some problems.

As the majority of reactor pressure vessels work in the temperature range from 300 to 350°C, the tests for strength properties at these temperatures are necessary, while it must be expected that yield point will decrease by 40 N/mm², and strength value by 70 to 100 N/mm² as compared to the values at room temperature. This requires that much attention is paid to heat treatment of forgings, rolled products or of the whole pressure vessel. Main type of heat treatment used is normalizing and tempering. With plates thicker than 100 mm accelerating cooling from normalizing temperature using a fine water shower or steam is often used. Such treatment results in refinement of the

structure, mainly in the middle of plate or forging, and helps also to increase its strength properties.

With low-alloyed plates, very often annealing within the temperature range $A_1 - A_3$ is being inserted between normalizing and tempering recently. This annealing results in further refinement of grains and improvement of plate's plastic properties. Normalizing temperatures vary depending on steel's chemical composition, mainly carbon content, within the temperature range between 830 to 930°C, and tempering temperatures between 600 to 650°C.

If heat treatment with partial austenitization is applied, then the temperature range between 740 to 780°C is chosen.

To obtain better homogeneity of plate, also double normalizing heat treatment can be applied.

In case of quenching and tempering, oil or water are used as a quenching medium, or possibly, forging or plates are quenched in press. With steels which are susceptible to tearing, surface inspection is carried out still during the process of cooling at the temperatures between 350 to 150°C. Owing to low content of impurities, the required plastic properties and homogeneity, as well as decreased carbon content, the weldability of reactor vessel steels is good. However, limiting conditions are required because of great thickness and shape intricacy of the vessel's body. Usually, automatic welding as e.g. gas shielded or submerged arc methods are regarded as most suitable. Electroslag welding is also used, if chemical composition

of plate permits it. Pre-heat temperatures range between 250 to 350°C; higher temperatures are not recommended due to the danger of hot tearing. Specific heat input is not limited, with some technologies /S-A, B-S/ however optimum temperatures are indicated /e.g. 30 to 35 kJ/ cm for S-A welding/. Post welding heat treatment usually consists of stress relief annealing in the temperature range from 580 to 660°C only. The temperature of weldment should not decrease between welding and annealing below 180 to 150°C. Post weld heat treatment is to:

1. Eliminate or reduce the level of internal stresses of the 1st order
2. Improve the microstructure of the HAZ and weld metal
3. Minimize the risk of delayed tearing
that

Owing to the fact heat treatment /annealing/ must be applied after each weld, some parts of the vessel may be exposed to the annealing effect below A_1 even for several tens hours. The welds as well as majority of ferritic steels change after such heat treatment their mechanical properties. It is therefore necessary with reactor steels to ensure their tempering stability.

The microstructure of reactor pressure vessel steels is ferrite-pearlitic or, in reactor steels with lower tensile strength /1st generation/, ferrite-carbide if they are used in as normalized condition. If they are Q and T their structure is ferrite-bainitic or decomposition /sorbitized/ mixture of carbides and ferrite with sporadic islands of pro-eutectoid ferrite. With alloyed steels /IInd generation/ no pro-eutectoid ferrite is usu-

ally found in the structure.

Owing to the fact that in majority of pressure vessels the internal surface is hard faced or plated with austenitic corrosion resistant material, high resistance of these steels to formation of inhomogeneities/rcheat cracks in particular/during surfacing is required.

Among the most important criteria for choice of material for reactor pressure vessel is its radiation stability. This is the reason why copper and cobalt content in the steel must be limited /Cu max 0.15%, Co max 0.02%. Especially in the case of alloyed steels it is difficult to keep the Co content on permissible level as this element is present as an accompanying element in many ferro-alloys.

G. 9% Ni steels

Recently, in power engineering also the steels of "Maraging" type are founding their application due to their good plastic properties at minus temperatures, good weldability and structural stability. In addition to 9% Ni content, they have low carbon content, adequate silicon and manganese content and reduced content of impurities. Chemical compositions and some properties are tabulated in Tab. 4.

Steels of this composition overcome the allotropic transformation $\gamma^r \rightarrow \alpha'$, while the temperatures M_s and M_f .

/same as the temperatures A_S and A_F at heating/ change according to the steel's chemical composition and its previous heat treatment, e.g.:

$$\begin{array}{ll} M_S = 420-440^{\circ}\text{C} & M_F = 320-350^{\circ}\text{C} \\ A_S = 550-580^{\circ}\text{C} & A_F = 600-740^{\circ}\text{C} \end{array}$$

Mechanical properties depend on the character of martensite /its sub-structure, density of lattice imperfections, degree of re-crystallization, morphology, presence of other phases/.

9%Ni steel ranks among maraging steels where the optimum properties can be attained only after certain changes in martensitic matrix. These changes are attained by combined treatment, its main component being heating to temperatures near to A_S or A_F . At this, partly the precipitation of hexagonal carbide or cementite in martensitic matrix, partly the re-crystallization of martensite or polygonization processes, and finally - this being the most important - the reverse reaction $\alpha' \rightarrow \beta' \rightarrow \alpha$ are taking place. Even after heating above the A_S temperature and repeated cooling to room temperature, structure of the alloy will be martensitic, but its sub-structure will already be changed. Further, at such treatment certain amount of residual austenite islands will remain in the martensitic matrix. An opinion prevails that this residual austenite /optimum amount of which does not overreach 10%/ contributes to good notch toughness. It is able to dissolve carbon, nitrogen and further elements /Ni, Mn, Cu/, so

that the neighbouring matrix is then non-aging. Very important is the temperature of tempering after quenching or normalizing. Most suitable is the temperature about 570°C , at which the residual austenite is stable up to the temperature below -200°C . Higher annealing temperatures cause the increased content of less carbon saturated austenite which decomposes at higher temperatures /above- 196°C / - e.g. at tempering at 790°C it decomposes already at 320°C . The structure of in this way treated material consists of tempered martensite /with ferritic sub-grains and small amount of carbides/, and stable austenite dispersed on grain and sub-grain boundaries.

The most common heat treatment of 9%Ni steel is double normalizing with subsequent tempering, but also quenching in water and tempering is used. In the latter case steel shows somewhat higher strength characteristics and, it seems, somewhat better notch toughness. Structure of the steel is similar in either case.

Material treated in this way gains at rather high strength characteristics also high toughness and excellent low-temperature properties. When the test temperature is decreased up to -200°C , yield and strength points increase proportionally to the temperature (Y.S. up to $1,000 \text{ N mm}^2$ and more, U.T.S. up to $1,300 \text{ N mm}^2$ and more/, ductility and contraction are changed inexpressively, and plasticity is consequently maintained. Equally good are also fatigue properties - e.g. at high-cycle loading the fatigue limit takes about 60% of the U.T.S. Notch toughness at -196°C is usually still higher than 50 J.

Captions

Fig. 1. Relation between yield point and transition temperature of notch toughness for low-carbon steels with different microstructures

Fig. 2. Relation between grain size and yield point of low-carbon steels with different structures

Fig. 3. Diagram of austenite decomposition under conditions of welding for conventional 2 1/4 % Cr 1 % Mo steel

Table 1. Brief review of heat resisting steels

Table 2. Steels for turbine rotors

Table 3. Chemical composition and some properties of steels intended for reactor pressure vessels

Table 4. Chemical composition and some properties of 9%Ni steels

Table 1

Brief review of heat resisting steels

Steel type	Chemical composition										Mechanical properties							
	C	Mn	Si	Cr	Ni	Mo	V	Ti	Nb	Al	Y.P. N.mm ⁻²	Y.P. N.mm ⁻²	U.T.S. N.mm ⁻²	R	U.T.S. N.mm ⁻²	U.T.S. N.mm ⁻²		
CrMoV	0,10	0,45	0,15	0,50	-	0,40	0,22	-	-	-	363	490	20	5P	500°C	580°C		
15128 ČSN	0,18	0,70	0,40	0,75	-	0,60	0,35	-	-	-	363	225	690	-	148	88	106	61
21/4Cr-1Mo	0,10	0,5	0,25	2,3	-	1,0	-	-	-	-	265	500°C	440	40	550°C	580°C		
10CrMo9.10, 15313	0,20	0,55	0,30	12	0,5	1,0	0,3	-	-	W	490	500°C	680	10	50	550°C	600°C	
2Cr12MoW	0,20	0,55	0,30	12	0,5	1,0	0,3	-	0,5	-	294	830	10	50	167	133	84	55
17.134	0,08	1,8	0,45	17,5	11,5	-	-	-	10xC	-	206	600°C	540	35	86	600°C	650°C	
08Cr19Ni9Nb	0,08	1,8	0,45	17,5	11,5	-	-	-	10xC	-	206	600°C	540	35	86	600°C	650°C	
AISI 347	0,08	1,3	0,45	16,5	13	-	-	-	10x	-	206	550°C	539	35	168	122	108	76
08Cr16Ni13Nb	0,08	1,3	0,45	16,5	13	-	-	-	10x	-	206	550°C	539	35	168	122	108	76
8 R 41	0,08	1,3	0,45	16,5	13	-	-	-	10x	-	206	550°C	539	35	168	122	108	76
06Cr17Ni13Mo2	0,06	max	max	16,0	11	2,0	-	-	-	-	205	550°C	500	35	158	110	100	68
AISI 316 H	0,10	2,0	0,75	18,0	14	3,0	-	-	-	-	205	550°C	500	35	158	110	100	68
17 483 ČSN	0,05	16,5	0,25	7,00	-	0,50	0,45	-	-	-	196	550°C	590	25	147	600°C	650°C	
08Cr2Mo1Nb1	0,08	0,45	0,25	2,3	0,6	1,0	-	-	1,0	-	245	500°C	440	-	550°C	-	-	-
03Cr18Ni11	max	max	max	17,0	10,0	-	-	-	-	-	175	600°C	490	-	550°C	650°C		
08Cr21Ni32Ti	0,030	2,0	1,0	20,0	12,5	-	-	-	-	-	175	600°C	490	-	550°C	650°C		
A1	max	max	max	19	30	-	-	0,15	-	0,15	166	600°C	480	-	650°C	750°C		
SANICRO 31	0,08	1,5	1,0	23	35	-	-	0,60	-	0,60	108	300°C	785	15	98	72	42	29
C5Cr13Ni16	max	0,8	0,4	13	6	/1/	-	-	-	-	590	300°C	785	15	98	72	42	29
2RW2	0,05	-	-	-	-	-	-	-	-	-	590	300°C	785	15	98	72	42	29

Table 2 Steels for turbine rotors

Steel designation	User	Chemical composition (% by wt.)								Y.P. N.mm ⁻²	U.T.S. N.mm ⁻²	δ ₅ %	R ₃ impact Jcm ²
		C	Mn	Si	Cr	Ni	Mo	V					
NiCrMo	Richardson West Garth	0,25	0,8	-	1,50 2,20	0,80 1,20	0,50 0,90						
2NiCrMoV	"	0,24	0,65	-	1,0	2,4	1,0	0,21 Nb=0,08					
561 S	Brown Boveri	0,18 0,25	0,25 0,80	0,10 0,40	1,20 2,00	0,90 1,10	0,50 0,80	max 0,05	600	900	min 15	70	
25Cr2Ni4MnFA	CNITTMAŠ	0,21 0,29	0,20 0,45	0,15 0,35	1,60 1,90	3,30 3,70	0,35 0,60	0,08 0,15					
34CrNi1A	Khar'kov's plant	0,30 0,38	0,40 0,70	0,17 0,37	0,90 1,20	0,40	0,40 0,55	-					
15 J20	IBZ KG Brno ČKD Praha	0,20 0,28	0,50 0,80	0,17 0,37	1,10 1,40	-	0,55 0,75	0,15 0,30	600	900	min 15	40	
Škoda N	Škoda, ČKD	0,24 0,30	0,40 0,60	0,17 0,37	1,40 1,70	2,80 3,20	0,50 0,60	0,03 0,08	750	900	18	98	
NK	ČKD	0,18 0,25	0,25 0,85	0,10 0,40	1,40 2,00	0,90 1,10	0,60 0,80	max 0,06	60 650	750 900	15	68	
16 221.6	I BZKG Brno	0,18 0,24	0,70 1,00	0,15 0,40	max 0,24	1,00 1,50	-	0,10 0,15	450	600	15	50	

Table 3

Chemical composition and some properties of steels intended for reactor pressure vessels

No. Generation	Steel	C	Mn	Si	P	S	Ni	Cr	Mo	V	Co	Heat treatment	Y.P. kg/mm ²	U.T.S. kg/mm ²
I	ASTM A2128	0,31	0,85 1,20	0,15 0,30	max. 0,035	max. 0,040						normalised annealed	270	490- 600
	ČSN 13 030-9	0,20	1,10 1,40	0,20 0,40	max. 0,020	max. 0,015	0,30 0,45	0,20			max. 0,01	-"-	220	440
3	ASTM A 302 B	max. 0,25	1,15 1,50	0,15 0,30	max. 0,035	max. 0,040			0,45 0,60			-"-	350	560 700
	A 350 - LF1	0,15	0,79	0,25	0,027	0,033	1,71	0,05		0,04		-"-		
5	ASTM A 533 A	max. 0,25	1,15 1,50	0,15 0,30	max. 0,035	max. 0,040			0,45 0,60			CT	350 490	
	ASTM A 542	max. 0,15	0,30 0,60	0,15 0,30	max. 0,035	max. 0,045		2,00 2,50	0,90 1,10			CT	600 700	740 810
7	ASTM A 543	max. 0,23	0,40 0,35	0,20 0,35	max. 0,020	max. 0,020	3,00 4,00	1,50 2,00	0,45 0,60			CT	600 700	740 810
	ČSN 15 219	0,06 0,10	1,30 1,80	0,15 0,30	max. 0,030	max. 0,025	max. 0,30	max. 0,85	0,55					
9	Škoda 15Ch2NF	0,08 0,15	0,4 0,7	0,17 0,37	max. 0,035	max. 0,035	max. 0,4	1,5 1,8	0,20 0,30	0,15 0,30	max. 0,01	CT	690 990	
	MOA	0,22	0,40	0,17	max.	max.	0,40	1,50	0,20	0,15	max.	CT	600	750
10	GOST 25Ch2N	0,29	0,70	0,37	0,035	0,035	0,70	1,80	0,30	0,30	0,01	CT	600	5000
	MOA	max.	max.	max.	max.	max.	17,0	4,6						
11	ASTM A 538 M	0,03	0,10	0,10	0,010	0,010	19,0	5,1			7,0 8,5	maraging	1620 1830	1680
	HY - 130T	max. 0,12	0,60 0,90	0,20 0,35	max. 0,010	max. 0,010	4,75 5,25	0,40 0,70	0,30 0,65	0,05 0,10		CT	910	
13	12 - 5 - 3N	max. 0,03	0,20 0,35		max. 0,010	max. 0,010	max. 12,0	max. 4,75	max. 3,0			maraging	1270	

Table 4

Chemical composition and some properties of 9% Ni steels.

Steel type	C	Mn	Si	Ni	P	S	Y.S. (N/mm ²)	U.T.S. (N/mm ²)
A 553 - 70 (ASTM)	0,13 max	0,9 max	0,15 0,30	8,5 9,5	0,040 max	0,040 max	517	690 830
X 8 Ni 9 (W.Germany)	0,10 max	0,8 max	0,15 0,30	8,5 9,5	0,030 max	0,030 max	490	640 840
KLN 9 (Japan)	0,13 max	0,9 max	0,15 0,30	8,5 9,5	0,025 max	0,030 max		
17 501 (Czechoslovakia)	0,10 max	0,30 0,80	0,15 0,35	8,5 10,0	0,030 max	0,030 max	490	640 830

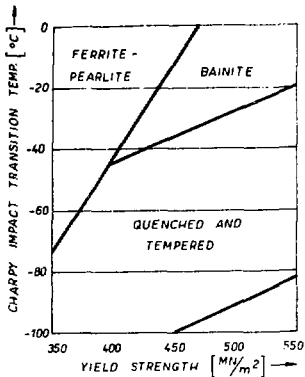


Fig. 1

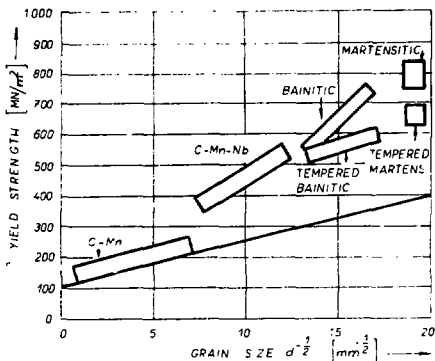


Fig. 2

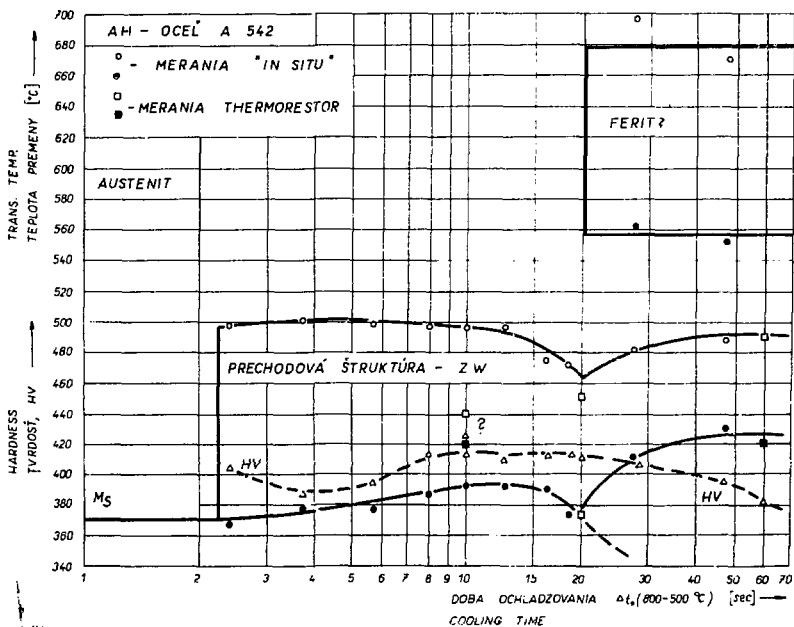


Fig. 3

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CONSUMABLES FOR SUBMERGED ARC WELDING IN POWER INDUSTRY

In submerged arc welding the welded joint quality depends on the correct choice of consumables i.e. the combination of flux and filler wire. The present paper deals with problematic of fluxes, as the flux to distinguish from filler wire, not only influences the metallurgical features of weld metal, but is the decisive factor for the technological proces of welding as well.

Whereas in case of the filler wire choice the general rule is holding to which in case of welding a given steel quality a wire of matching chemistry is chosen, this is not the case with the flux choice. The flux as such has a far broader influence and its aplicability is generally not restricted on a certain steel quality. The same flux can be applied in welding of a whole range of steel qualities, this in many cases of alloyed as well as unalloyed steels.

The scale of steels applied in welding of products for power industry is very broad. Here unalloyed, micro-alloyed, low-alloyed steels for service temperatures up to 450°C as well as high-alloyed austenitic steels of CrNi-, MnCr-type can be found. Besides that the considerable part is represented by building up high-alloyed deposites by the strip electrode, above all in the fabrication of nuclear-power station components. Owing to such broadness of this problematic we held for useful to treat the fluxes from a general viow.

1. Influence of the flux basicity grade.

The basicity grade of the fluxes is by most of the authors clasified as the decisive factor influencing the metallurgical features of weld metals. To express the flux basicity grade a great scale of relations has been elaborated. For the most suitable I hold the one elaborated by Tuliani Boniszewsky and Eaton /1/ namely:

$$B = \frac{\text{CaO} + \text{MgO} + \text{BaO} + \text{SrO} + \text{K}_2\text{O} + \text{Li}_2\text{O} + \text{CaF}_2 + \frac{1}{2} \text{MnO-FeO}}{\text{SiO}_2 + \frac{1}{2} \text{Al}_2\text{O}_3 + \text{TiO}_2 + \text{ZrO}_2}$$

The contents of oxide components is given in % .

This relation is the expression for reducing and oxidizing features of fluxes. Its use makes it possible to determine e.g. the relation between the basicity grade B and the content of silicon in the weld metal. This relation for determining the basicity grade has been applied for the flux classification according to IIW /2/.

Recently a nex law for the metallurgical flux characteristic has been introduced by Zeke /3/. This relation is of a similar meaning as the flux basicity-grade. To distinguish of the previous relations based on the molecular theory of the slug systems structural characteristic Zeke uses the ion-theory approach. This new metallurgical characteristic is designated as $\text{no}_{\text{CH}}^{2-}$ and reads:

$$\text{no}_{\text{CH}}^{2-} + m\text{RO} + m\text{R}_2\text{O} = /2m\text{SiO}_2 + 2m \text{TiO}_2 + m\text{Al}_2\text{O}_3/ \text{ where:}$$

mRO is the sum of the oxide-moles of doublevalent elements and nO_{CH}^{2-} means the free oxide gramions number O^{2-} missing for the completion of the whole SiO_2 , TiO_2 and Al_2O_3 content into complex anions SiO_4^{4-} , TiO_4^{4-} , and AlO_2^- . The lower is the absolute value of the nO_{CH}^{2-} -expression, the higher is the basicity of the flux. The metallurgical characteristic makes it possible to express quantitatively by means of the so called silicon reduction factor $/KIR_{Si}/$, or the manganese reduction factor $/KIR_{Mn}/$ the silicon or manganese reaction respectively, according to the following expressions:

$$KIR_{Si} = nO_{CH}^{2-} \cdot NSiO_2 \cdot /1 - \sum N MnO/$$

where: $NSiO_2$ represents the molar fraction SiO_2 ,

$NMnO$ represents the molar fraction MnO and

$$KIR_{Mn} = NMnO \cdot /1 - KIR_{Si}/ .$$

The flux basicity generally influences favourably the mechanical properties of weld metal /4/. Encreasing basicity decreases the weld metal Si-content. This is connected with the reduction of nonmetallic silicate inclusions and herewith also of the oxygen contents in weld metal. Tuliani /4/ e.g. reports that 0,1 Si reduced from of flux imports into the weld metal 0,01% O_2 . Increasing basicity influences favourably the S-content /4,5/.

The basicity effect is often put into connection with the favourable influence of flux on the fine-grain structure formation /6/. This is attributed to the flux

physical properties. The basic fluxes give the so called "short slug" in general, i.e. slug with a narrow solidification interval. The narrow flux solidification interval is a favourable contribution to the fine-grain microstructure.

The mentioned favourable effect of the flux basicity predetermines the basic fluxes for welding of uneasier weldable low- and/or high-alloy steels, if it is difficult to attain the required plastic properties namely the notch toughness values. As just welding of the power industry equipment concerns the more pretentious steel qualities the basic fluxes are finding in this sphere broad application. On the other hand the use of acid fluxes which in practice are represented by the so-called high silicon and high manganese types comes into question only in unalloyed structural, eventually microalloyed steels.

2. Influence of some main flux components.

Although the main criterion from the point of metallurgical influence is the flux basicity we take it necessary to point to the effect of some main flux components governing mainly the transition of some elements into the weld metal. The influence of the components on the transition and/or melting loss of alloying elements is essential especially in welding of some high-alloy CrNi steels as well as strip depositing by electrodes of this type which occurs above

all in the fabrication of nuclear power stations.

The influence of some basic components in flux on the C, Si, Mn, Cr, Ni content in weld metal can be assessed from experimental results acquired in the framework of the development of the basic flux VÚZ-4 BaL /8f.

To assess the transition of the mentioned elements, five runs deposits were built-up with the 18/15 Cr/Ni type wire whereby the fifth deposit layer has been analysed. The chemistry of the applied wire was 0.022 C, 1.56 Mn, 0.24 Si, 19.87 Cr 1513 Ni /%/. In the experimental fluxes the SiO_2 , CaO, Al_2O_3 , MnO and CaF has been varied.

The SiO_2 influence.

The SiO_2 content has been dosed into the dry flux mixture such as the separate trial fluxes would contain 2 %, 4 %, 6 %, 8 %, 10 %, 12 % or 15 % SiO_2 whereby the remaining components ratio remained unchanged, as in the base flux.

The influence of the changing SiO_2 content in flux on the studied elements /except C/ transition is processed in diagram on Fig. 1. It became evident that with melted basic type fluxes carbonisation does not take place and the studied flux components practically do not influence the C- content in the deposit. From Fig. 1. it is evident that with increasing SiO_2 content an expressive increase in Si transition takes

place. On the other hand the Mn transition is diminishing and with SiO_2 -content in flux higher than 6 % the Mn-content in deposit is lesser than the original Mn-content in wire. That means that Mn due to the manganese-silicon balance is oxidizing and transiting into slag. The SiO_2 increment affects unfavourably the Cr melting loss which although slightly is nevertheless continually increasing.

The Ni-content remains practically unchanged.

The CaO-influence.

In trial fluxes the CaO-content has been dosed in quantities of 6, 8, 10, 12, 14, 16, 18, 22 and 24 %. The established results are plotted in diagram Fig. 2. CaO being the basic component influences the Si-transition into deposit reversely than SiO_2 whereby, however, the CaO influence is not so expressive as with SiO_2 . With a CaO-change of 18 %, Si-content in deposit changed only 0.26 %^{by}, whereas with a 16%-change of SiO_2 the established change of Si in deposit was 0.47 %. By adding CaO the Mn-content in deposit is increasing. The Cr-melt loss is unfavourably affected by the increasing CaO-content. Like with SiO_2 the CaO-change does not influence the Ni-content.

The Al_2O_3 -influence.

The composition of the studied elements in deposits built-up with fluxes, the Al_2O_3 -content of which varied

from 32.5 to 52.5 % is on Fig. 5. The Al_2O_3 -increment is causing a slight increase in the Si-transition and increase in the Cr-melt loss. The influence of Al_2O_3 on the further two studied elements is on the whole not apparent.

Influence of the separate flux components can be summarized as follows:

The alloying elements melting loss especially of Cr is favourably influenced by fluorides as well as by the CaO and MgO basic oxides. The melting loss is on the other hand unfavourably influenced by SiO_2 , MnO and also by Al_2O_3 .

The transition of Si which can have unfavourable effect on plastic properties, eventually influence also crackability is above all governed by the SiO_2 content in the flux.

From the above it ensues that for welding and/or surfacing of high alloy steels the mostly favourable would be the non silicon fluxes consisting of fluorides and oxides of the mostly negative thermodynamic potential represented mainly by CaO and MgO.

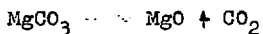
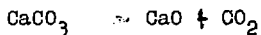
The non silicon fluxes do, however, in practice not exist, as with such fluxes no suitable operative and/or shaping properties of welds could be attained, From the point of shaping it is essential that fluxes, mainly in case of surfacing by strip electrode on which from the point of operative and shaping properties high demands are claimed, should contain certain quantity of SiO_2 .

This is important also for securing freedom from porosity in the deposit. For that reason the realistic fluxes contain minimum from 6 to 8 % of SiO_2 .

3. The problematic of melted and unmelted fluxes.

According to the method of fabrication the fluxes are classified principally into two categories, namely the melted and the unmelted ones. The unmelted fluxes can be fabricated by various technologies. To-day, however, definitely the production technology of the agglomerated fluxes is predominating.

When comparing melted and unmelted fluxes, as the main merit of the melted ones the better operative and/or shaping properties could be held. The reason for that is the fact that with melted fluxes some components, especially carbonates are decomposing already during their fabrication - melting process - so that the flux is degasified. In unmelted fluxes these components remain in their original form, e.g. CaCO_3 or MgCO_3 . In course of welding these carbonates dissociate according to reactions:



The arisen carbon dioxide has then unfavourable effect on the weld process stability and/or deposit shaping.

The further merit of the melted fluxes is the fact that they do not contain any hygroscopic components, do not show aptness to wettability and therefore are showing lower ^{susceptibility} to porosity. The melted fluxes can manifest themselves from the point of S and P content in the weld metal. This is the consequence of the fact, that during the melting process a partial reduction in sulphur, eventually phosphorus content in the flux is achieved and this influences then secondarily favourable also the S and P content in weld metal.

However from the metallurgical point the melted fluxes offer less possibilities than the unmelted ones. Besides Si and Mn, the melted fluxes are giving only a slight possibility to alloy the weld metal by further additions as the alloying elements, or deoxidizing substances cannot remain in the flux in a non reacted form as they already oxide in course of the melting process during the fabrication.

The unmelted fluxes offer in general more favourable conditions for securing higher intensity of the reaction between slag and metal. By means of unmelted fluxes it is e.g. possible to alloy the weld metal by different elements and this in a relatively wide range as well as to secure a good deoxidation of the weld metal.

Owing to these merits, the unmelted fluxes are applied presently mainly in welding the difficultly weldable low alloy and/or high strength steels as well

as steels destined for low temperature service.

For illustration we give in Fig. 6 the plot of notch toughness of the weld metal deposited with help of the agglomerated VÚZ-N3AlBa flux /6/. For the sake of comparison the plot includes also courses of the notch toughness mean values which were established under similar conditions with the selected melted fluxes VÚZ-2Ba, VÚZ-4BaF and VÚZ-34Mn.

4. Conclusion

With respect to the broad range of steels coming into consideration for welding in the power industry there is the possibility to turn to account fluxes of different type. The flux choice can be governed by the following principles:

For welding unalloyed structural steels the high manganese and high silicon fluxes are to be used.

For welding of low alloy and/or high alloy steels the application of basic type flux is favourable.

For welding difficultly weldable alloy steels the unmelted basic fluxes are especially favourable, this due to their higher metallurgical activity.

For strip electrode welding and/or surfacing of high alloy Cr/Ni-type steels, with regard to melt losses of the alloying additions it is suitable to produce fluxes of high fluorine components content, containing at the same time CaO and MgO.

The choice of the fillet wire is generally governed by the principle, that the wire should be of the same

chemistry as the welded steel, whereby it can be improved by some alloying additions and by of a lower impurity grade than the parent metal.

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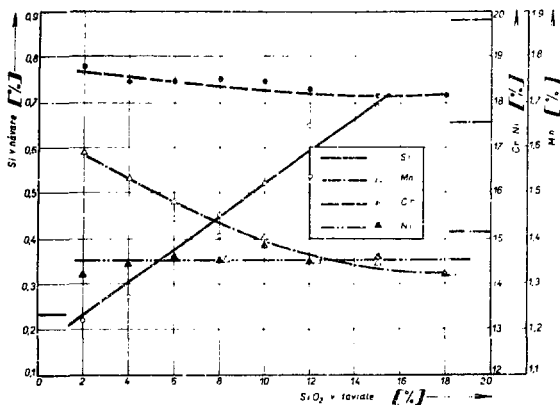


Fig. 1

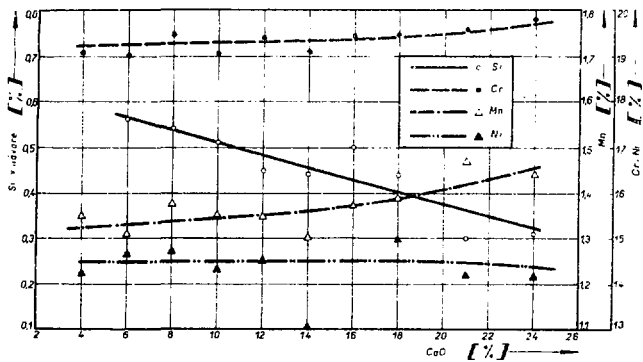


Fig. 2

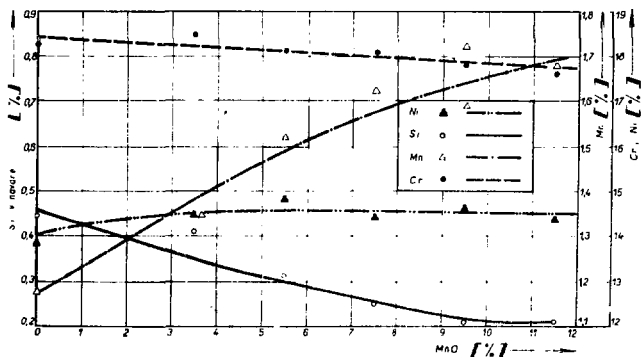


Fig. 3

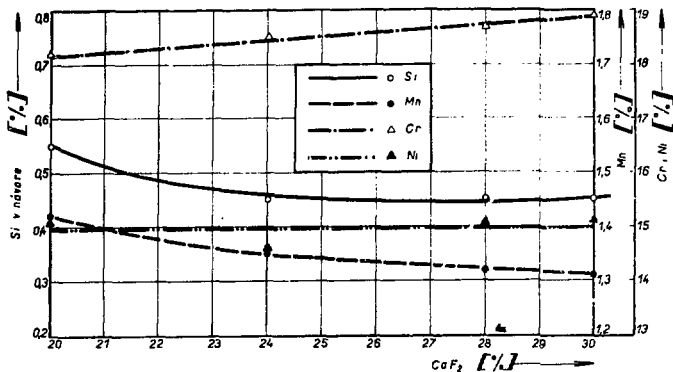


Fig. 4

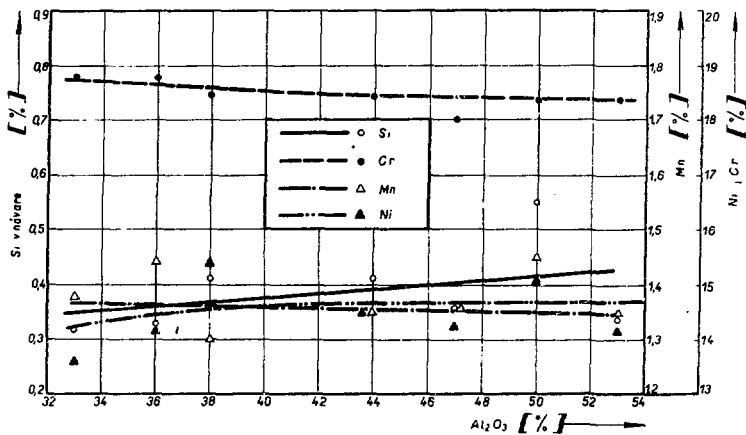


Fig. 5

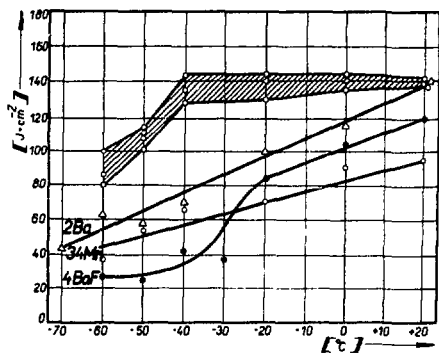


Fig. 6