Problems of Materials Processing and Joining

I. Hrivnak[+]

The recent developments in metallurgy result in new type of steels with excellent toughness and higher strengths. The application of thermomechanical rolling and/or accelerated cooling makes possible further improvements in steel quality. When welding very clean high strength structural steels a deterioration of toughness can take place in the underbead zone due to grain coarsening and transformation embrittlement. The brittle fracture tests like CTOD can reveal also the effect of local brittle zone. There are several mechanisms of local embrittlement, namely in multipass welds. The main among them is connected with the grain coarsening.

Modern structural steels have improved weldability, in general. They are more suitable for application of advanced welding processes.

The main motivation for the remarkable development of never steels and other structural materials is the demand of industry for materials able to support high service stresses and to withstand brittle failure under extreme loading condition. The need arises therefore for high strength steels. The way to enhance the strength of steel produced by conventional rolling or normalizing treatment is by increasing alloy elements. Higher strength and to

[+] Department of Material Science Faculty of Metallurgy
Technical University Košice, Czechoslovakia

some extent also an improvement of toughness is impaired by quenching and tempering. Thermomechanical controlled process (TMCP), has become remarkably advanced using accelerated cooling and direct quenching. In TMCP, pre-rolling is carried out as in the case of normal rolling, possibly at some what lower temperatures and finish rolling at substantially lower temperatures, which permit little if any recrystallization of the austenite. TMCP therefore allows to compensate the lowered carbon content by precipitation hardening and grain refinement, while the accelerated cooling allows to control the final microstructural transformations.

Improved quality of modern steels arises also from the treatment of the liquid metal and in conducting the solidification. Desiliconization can be carried out by brasting of iron oxide on hot liquid metal in runner while desulphurization is supported by mechanical stirring of hot metal in ladle, dephosphorization by low phosphorous BOF slag and final desulphurization and inclusion shape control by introduction of processes which lead to TiN precipitates smaller than 0.05 μm and small particles of Ca (O,S) and REM (O,S) (Rare earth metals). The TiN precipitates, complex precipitates of REM-B, TiN-MnS-$Fe_{23}(CB)_6$ and Ti-oxides are also adopted as nucleation sites for ferrite and bainite inside the coarse austenite grains.

As it was mentioned TiN particles are used to inhibit austenite grains from coarsening. However, TiN-precipitates dissolve at highest sub-solidus temperatures and therefore the inhibition effect of these particles is weak. More suitable from this point of view are the particles of Ca (O,S), or REM (O,S). The amounts of Ca and REM are usually limited to 50 ppm so that the precipitates are controlled within 2-3 μm in size and act as inhibitors.

Concerning the precipitates as nucleation sites of ferrite Ti-oxide is more effective than complex ones of TiN and Ti-B in the coarse-grained region affected over 1400 °C.

Modern mettallurgical processes are improving steel veldability. Lovering the carbon content in the steel and reducing the amount of alloying elements like Cr, Mo, reduces also the crackability of the steel during velding. But there are still problems to get satisfactory toughness at low temperatures in some areas of velds.

Cold cracking

Cold cracks or hydrogen induced cracks or delayed cracks are cracks in veld metal or HAZ (heat affected zone) which are forming in the final stage of veld production or after the velds have been completed. In general, the view is accepted that cold cracks are caused by the mutual action of three factors, viz:

- The presence of hydrogen in the veld.
- The presence of a microstructure sensitive to hydrogen.
- The presence of tensile (contractive) residual stresses.

The hydrogen initially introduced and dissolved in the veld metal tends suddenly to diffuse through the bond face to the HAZ when the transformation from austenite to ferrite or hard decomposition structure takes place, while the HAZ is still in austenite state.

The avoidance of cold cracks is very important, and many attempts vere described [1-6] to specific the cold cracking behaviour of steel by a suitable parameter like carbon equivalent. Carbon equivalent formulae (CE) vere derived for a number of purposes, as hardenability, cold cracking sensitivity or evaluation of service properties for which a correlation is expected with the hardness, like sulphide stress corrosion cracking. It is generally accepted that the well known IIW CE

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

may be used for carbon steels having more than 0.18 % C, or in the case of large heat inputs, in which the $t_{8/5}$ cooling time is longer than about 12 s. The Ito and Bessyo [1] P_{CM} formula

$$P_{CM} = C + \frac{Si}{30} + \frac{Mn + Cu + Cr}{20} + \frac{Ni}{60} + \frac{Mo}{15} + \frac{V}{10} + 5\,B$$

may be prefered for lower carbon content steels and in the case of rapid cooling, $t_{8/5}$ less than 6 s. The Yurioka [3,5] formula

$$CEN = C + A\,(C)\,\left(\frac{Si}{24} + \frac{Mn}{6} + \frac{Cu}{15} + \frac{Ni}{20} + \frac{Cr+Mo+Nb+V}{5} + 5\,B\right)$$

where $A(C) = 0.75 + 0.25 \tanh\,[20\,(C-0.12)]$ gives good evaluation for structural steels. To avoid cold cracking measures must be taken like to use the proper preheating or to controll the amount of hydrogen in weld. The necessary preheating temperature can be either calculated or determined on the bases of cold cracking tests. The occurence of hard decomposition microstructure is one of the most important factors in hydrogen cracking. However the susceptibility to cracking is influenced also by local concentration of hydrogen and local stress state. Prediction of the local state of hydrogen and stress in weld is very difficult. Even there are some models for calculating of the hydrogen behaviour in welds and necessary preheating temperature, it is practical to use methods based on empirical data. Most of the methods are based on the results of cold cracking tests with consideration of carbon equivalent of steel, weld metal hydrogen content, plate thickness and heat input. An example of the critical preheating temperature determination for butt welds made of properly dried basic electrodes for steels with various CEN values is shown in Fig.1 [5]. Uwer and Höhne [7] made the following proposal for calculation the preheating temperature:

$$T\ (^{o}C) = 700\ CET + 160\ \tan h\ \frac{d}{35} + 62\ HD^{0.36} + (53\ CET - 32)\ Q - 330$$

where $CET = C + \frac{Mn+Mo}{10} + \frac{Cr+Cu}{20} + \frac{Ni}{40}$

d is the plate thickness (mm)

HD is the amount of the diffusable hydrogen (in $cm^3/10$ g)

Q is the weld heat input (kJ/mm)

The effect of HAZ microstructure expressed by the hardness of HAZ was evaluated by Hart and Harrison [6]. Combining the result of CTS, bead-on-plate and implant tests (for the average value of HD = 10 ml/100 g) they supposed the maximum admissible hardness (from the point of view of cold cracking) as

$$HV = 283.3 + 668.1\ (C + \frac{Mn}{42} + \frac{V}{4} + \frac{Mo}{24})$$

and the corresponding critical cooling time

$$t_{8/5}\ (crit) = 3.7 \times (C + \frac{Mn}{13} + \frac{V}{6} + \frac{Ni}{40} + \frac{Mo}{10}) - 0.31$$

With modern structural steels produced by TMP in which the content of carbon and other alloying elements is reduced drastically, the transformation $\gamma \rightarrow \alpha$ in the HAZ starts earlier as the same transformation in weld metal. That´s why cold cracking is not so much an incident affecting the steel HAZ but a danger for the weld metal. Therefore selections for the weld consumables need a greater attention.

Lamelar tearing

High transient or residual tensile stress developed during welding in direction perpendicular to the surface may cause tearing within the steel. This tearing is usually associated with sulphide or oxide inclusions in the steel. The sensitivity to tearing may be increased by the presence of hydrogen. The principal factors affecting lamellar tearing are

- plasticity properties of steel plate in the through-thickness direction;
- mode of welding and parameters used;
- construction of the welded node.

The most important role in preventing tearing is played by the cleanliness of steel, inclusion (sulphide) surface, but also by oxide clusters, which may help to depress steel ductility in the through thickness direction. Modern structural steels, exhibiting high metallurgical cleanliness are less prone to delamination. Sulphide-shape control is also an avoiding treatment in preventing lamellar tearing.

Toughness properties in the heat-affected zone

The transformation embrittlement with the grain coarsening is an very important feature when evaluating fracture toughness of HAZ and are dependent to a great extent on welding conditions, in particular, heat input.

In most of the structural steels the grain start to coarsen when the temperature is exceeding ~ 1150 °C. Microalloyed steels are more resistant to grain coarsening, up to ~ 1350 °C. But above this temperature the grains start to coarsen very rapidly. The coarsening is inhibited by dissolution of second phase particles

- namely carbides and nitrides. Very clean steels are more prone to grain coarsening. Therefore the introduction of second phase particles into their microstructure is essential.

During the cooling period of the welding cycle coarse austenite grains start to decompose. The coarser are the grains the slover is the kinetics of austenite decomposition and hard decomposition structures like martensite or lower bainite prevail. In order to achieve adeqaute toughness properties, it is necessary to make some limitation concerning the weld heat input. Low heat input reduces the dwell time in which the underbead zone is kept at high temperatures and consequently also reduces the width of the coarse grained zone. At the same welding conditions, modern structural steels like TMP, are less sensitive to grain growth and due to their lover carbon and alloying elements content, they are less sensitive to martensitic transformation. The result is a tough bainite structure in the HAZ. Fig.2 is an example of impact transition curves of the underbead zone of TMP 355 MPa Yield strength steel using different heat input at welding [5].

Many modern structural steels having high strength and excelent toughness of the parent material exhibit a rapid drop of toughness in the HAZ when milder welding cycles are used. At slover cooling rates, the microstructure of the HAZ changes from martenzite to lower bainite, upper bainite and finally, to acicular and proeutectoid ferrite. The granular or upper bainite is composed of side plate structure with carbide colonies and for Martensite/ Austenite (M-A) islands in between depending on alloy content and heat input. The retained austenite in the M-A in milder decomposition structures can be partially transformed into pearlite if very low cooling rates are used. With higher cooling rates the retained austenite is partially transformed to martensite. When the transformation starts at higher temperatures (e.g. 450 °C) the microstructure of martensite is lath-like. If the carbon content of austenite is considerably higher (even above the eutectoid one), the Ms temperature is decreasing and the martensite laths exhibite internal twinning. Fig.3 shows the occurence of M-A constituent in the simulated underbead zone of HT 80 QT steel. The presence of retained austenite in M-A constituent is considered very harmfull for the toughness. Its volume fraction increases with increasing cooling time, $t_{8/15}$ as a consequence of carbon enrichment resulting from the increment of diffusion time, and usually is close to 2-6 %. According to Ikava et al. [8] retained austenite promotes initiation and propagation of the fracture.

Still very much is discussed the role of precipitates on the weldability and toughness of welds.

Well known and well understood is the effect of precipitates in the base metal of microalloyed steels. Addition of small portion of Al, Ti, Nb, V to steel makes possible to get finer grains. Further on nitrides or carbonitrides combine the interstitial nitrogen making so the steel less sensitive to ageing. A small excess of Al and Ti above the stechiometric values of nitrides is normally required to neutralize all free nitrogen in the weld metal, because oxidation of these elements and the formation of non-metallic inclusions reduces their free solute concentrations. Small precipitates (Fig.4) can also contribute to the strength of the steel. Anyhow, precipitation hardening has concenquences in precipitation embrittlement.

<u>Local brittle zone</u>

In multipass welds, the microstructure of the HAZ formed by the first pass may be changed by the subsequent weld passes. The coarse grained structure of underbead zone may be tempered by the second thermal cycle. If the steel is sensitive to stress relieving embrittlement caused by segregation of some impurity elements (P. S. Sb. As....) at prior grain boundaries. LBZ at this zone can be detected by CTOD tests. If the maximum temperature of the second velding cycle exceeds Ac_1 (intercritically reheated coarse grained zone), a precipitation of bainite (troostite) island at these boundaries causes an other type of embrittlement (LBZ).

So the LBZ in multipass veld can have various forms:

- coarsening of grains in underbead zone
- transformation embrittlement in this zone (martensite);
- the occurence of M-A constituent in granullar bainite. vhen using mild velding cycles;
- precipitation of bainite (troostite) islands in intercrytically reheated coarse grained zone;
- stress relieving embrittlement in subcritically reheated coarse grained zone ;
- and. finally. recrystallization of underbead zone vhich results in ferrite microstructure (Fig.5).

The LBZ is more sensitive in the CTOD evaluation tests. The lovering of the CTOD value belov 0.05 mm is observed if the LBZ is only 0.5 mm in size.

Conclusions

By introducing never and advanced velding processes materials and steels vhich are difficult to veld can be velded easily. It is therefore necessary to study the veldability of such steels. The use of computers and microprocessors make possible to keep the velding parameters very precisely. Therefore optimization of velding parameters is needed.

In spite of the achievements in metallurgy never and novel steels and structural materials are put on the market. In these steels strength is enhanced through the refinement of ferrite and bainite structures vithout increasing chemical composition; TMCP enables the reduction of CE in comparison vith normalizing.

Attention is paid in particular, to the structures that are used at lov temperatures or are dynamically loaded. In these structures brittle or fatique cracking can initiate at localized areas of velds.

<u>L i t e r a t u r e:</u>

1. Ito, Y., Bessyo, K.: Veldability formula of high strength steels. Doc. IIV-IX-576-68

2. de Meester, B.: Velding in the vorld, 1990, 28, No.3/4, p.48-54

3. Yurioka, N. et al.: Metal Construction, April, 1987, p.217R-223R

4. Terasaki, T. et al.: Chemical composition and velding procedure to avoid cold cracking. Proc. of the Int.Conf. JOM-2, Helsingor, April 15-18, 1984, p.381-386

5. Yurioka, N.: Veldability of modern high strength steels. Proc. of the First Japan - US symposium on Advances in Velding Metallurgy, Yokohama, June 12-13, 1990, p.27-58

6. Hart, P., Harrison, P.: Compositional parameters for HAZ Cracking and Hardening in CMn Steels. Doc. IIV-IX-1440-86

7. Uver, D., Höhne, H.: Characterization of the cold cracking behaviour of steels during velding. Doc. IIV-IX-1630-91

8. Ikava, H. et al.: J. of Velding (Japan), 7, 1977, p.396

9. Rauch, R., Schütz, H.: Advantages of TM-Steels in Velding Fabrication. Doc. IIV-IX-1647-91.

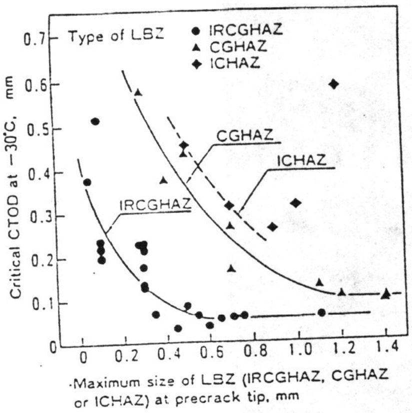

Fig. 1. Determination of the critical preheating temperature for butt welds with various CEN values [5].

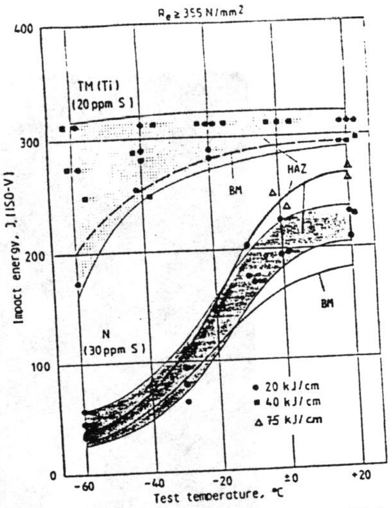

Fig. 2. The impact transition curves of the underbead zone of TMP 355 MPa yield strength steel in dependence to weld heat input [9].

Fig. 3. The M-A constituent in underbead zone of HT 80 steel. This foil. Transmission electron microscope.

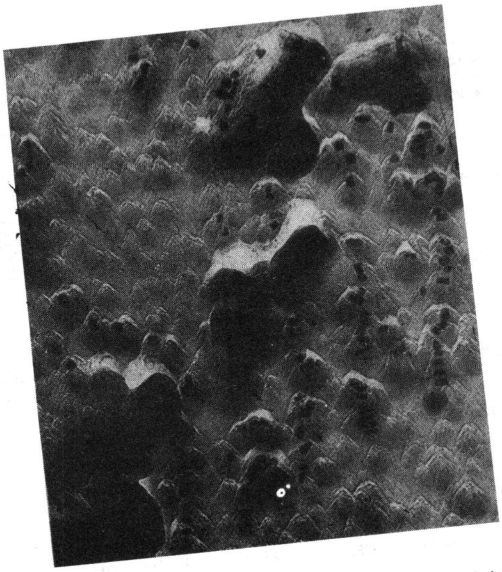

Fig. 4. Very fine precipitates (MX) in high strength micro-alloyed steel. Carbon extraction replica.

Fig. 5. Recrystallization of ferrite in bond face of submerged arc welded joint on X 22 steel. 200x.