MICROSTRUCTURE AND METALLURGY INVESTIGATION OF THE HOT DUCTILITY BEHAVIOUR OF STEELS

Pavliska J.[1], Jonšta Z.[1], Mazancová E.[2], Mazanec K.[1]

[1] Institute of Materials Engineering, Technical University Ostrava,

708 33 Ostrava-Poruba, Czech Republic

[2] Nová Huť a.s., Research and Testing Institute, 707 02 Ostrava-Kunčice, Czech Republic

MIKROSTRUKTURNÍ A METALURGICKÝ VÝZKUM TAŽNOSTI OCELÍ ZA TEPLA

Pavliska J.[1], Jonšta Z.[1], Mazancová E.[2], Mazanec K.[1]

[1] Katedra materiálového inženýrství, VŠB-TU Ostrava, 708 33 Ostrava-Poruba,

Česká republika

[2] Nová Huť a.s., Výzkumný a zkušební ústav, 707 02 Ostrava-Kunčice, Česká republika

Abstrakt

Práce je věnována studiu vlivu mikrostruktury a metalurgických parametrů ocelí na dosahovanou úroveň tažnosti za tepla v kritickém teplotním intervalu. Za těchto podmínek mikrostruktura oceli je tvořena směsí feritu a austenitu a deformace je lokalizována přednostně na hranice zrn austenitu. Deformace je koncentrována v intergranulárních feritických pásech. Její charakteristiky jsou diskutovány jednak z hlediska morfologie feritu, jednak z hlediska objemového podílu feritu ve struktuře. Současně je analyzováno působení reziduálních prvků, Cu, Sn, a Sb a jejich vliv na modifikaci dosahované tažnosti za tepla. Hodnocení jak podmínek vedoucích v oceli k povrchovému obohacení reziduálními prvky, tak důsledků tohto obohacení na rozvoj vysokoteplotní křehkosti jsou zahrnuty do předloženého řešení. Je navržen mechanismus příznivého vlivu a působení Ni.

Abstract

The work is devoted to the study of microstructure and metallurgical parameters on the level of hot ductility in the critical temperature range. Under this condition the steel microstructure consists of the ferrite and austenite mixture and the deformation is realised at the austenite grain boundaries. The characteristics of localised deformation in intergranular ferrite layers are discussed partly from view point of ferrite morphology, partly from view point of ferrite volume fraction. Simultaneously, the action of the residual elements (Cu, Sn, Sb) is analysed and their influence on the modification of attained level of hot ductility is discussed. The conditions leading in steel to the surface enrichment of these elements and the evaluation of its consequences on the development of hot ductility shortness are included in the presented solution. The beneficial effect of the Ni-addition and proposed mechanism of its action are analysed.

Key words: residual element, transverse cracks, localised deformation, hot ductility, intergranular ferrite, acicular ferrite

1. Introduction

The obtained level of hot ductility of the cast low-carbon steels plays a very important role during their deformation realised in austenitic and/or in complex ferritic and austenitic microstructure. The study is devoted to the elucidation of steel response to the localised deformation at the austenite grain boundary or in the softer intergranularly precipitated ferrite in heterogeneous ferritic/ austenitic matrix. These conditions correspond to the straightening operation of the continuously cast (CC) products [1]. The temperature of this operation falls in the range of 700 - 1100°C, which is known as temperature range corresponding to the formation of the hot ductility trough [2]. Usually, this behaviour is evaluated using the isothermal tensile tests that are performed at strain rate around 10^{-3} to 10^{-4} s^{-1}. This level of strain rate is chosen to be the same as that applied during the straightening operation of CC products [2].

2. Theoretical background

The appearance of the hot ductility shortness by tensile testing and the corresponding susceptibility to the cracks formation have the immediate relation to the ferrite morphology and the ferrite volume fraction in the investigated steel type. These microstructure characteristics are linked to the realised phase transformation process of austenite into ferrite that is running in cast steel during its straightening operation [3, 4]. Besides the above discussed microstructure effects, the presence of some residual elements in steel and their influence on a higher occurrence of cracks resulting from the detrimental behaviour of these elements on the attained tensile properties at above mentioned critical temperature range are taken into consideration.

The critical levels of residual elements as Cu, Sn, Sb and their unfavourable surface activity play a very important role by influencing of surface quality of steel products. These elements encourage cracks initiation owing to the weakening effect resulting from local matrix enrichment of the mentioned detrimental elements on the mechanical metallurgy properties of steel in question. Simultaneously, the "remedy" effects of Ni and Si are analysed. The influence of Cu > 0,15% is deleterious to the hot tensile ductility and this response can be intensified under the superposed action of Sn and Sb.

The aim of this work is to contribute as to the elucidation of some residual elements effects as to the analysis of embrittling influence of the microstructure parameters (the ferrite morphology) on the hot ductility shortness of steel in question and on the mechanisms of realised cracking processes.

3. Relation between ferrite properties and hot ductility shortness

The relation between the ferrite morphology and the susceptibility e. g. of CC steel products to the transverse cracks formation were discussed in whole range of works [5-7]. Through the laboratory studies, transverse cracking is linked to the low hot ductility values of tensile specimens tested after appropriate phase transformations in steel matrix during the cooling process. The chemical composition of steels plays an important role in their physical metallurgy behaviour, in particular from view point of austenite decomposition parameters, inclusive of the ferrite formation probability by deformation induced mechanism [8]. In carbon and/or in C-Mn steel, the appearance of the low ductility trough is related to the austenite transformation.

The ferrite formed in steel is softer than austenite at a given temperature. This is because ferrite has a higher stacking fault energy than austenite. Therefore, the ferrite recovers more rapidly.

The result of this property is that when small amounts of ferrite are present in an austenite matrix, any strain applied to the material is concentrated in softer ferrite regions.

The ferrite is therefore subjected to a deformation substantially higher than is its level imposed on the steel as a whole and exceeds this value several times [9]. The localised deformation is up to 20 times increased, approximately, in comparison with the global deformation level.

The realised analysis shows, the minimum of the hot ductility value can be found for ferrite volume fraction of about 5%. It should be emphasized that the above considered conception is only acceptable if the ferrite covers the entire austenite grain boundaries. For volume fraction of ferrite which is smaller than the proposed limit of 5% approximately, the discontinuities of intergranular ferrite layers are detected and the considered detrimental effect of these ferrite films is suppressed. On the contrary, if the ferrite volume fraction is higher than 30% then the strain localisation is less significant and this ferrite volume fraction becomes less detrimental from view point of the attained hot ductility level.

Failure in uniaxial tension test performed in the dual-phase region, simulating as closely as possible the conditions pertaining to the unbending operation during CC process, occurs by void linkage along the thin ferrite film decorating the austenite grain boundaries. The basic effect causing this low ductility trough formation is localised plastic deformation at the austenite grain boundaries if the intergranular sliding mechanism is dominant in realised deformation process. Besides this process, it is necessary also to take into consideration the participation of precipitated non-metallic inclusions (e.g. sulphides) and the physical metallurgy parameters of austenite decomposition into ferrite during the steel cooling from the solidification temperature [10].

A very interesting finding of the performed investigation is that increasing the strain rate improves the attained hot ductility level. The ductility trough dimensions are reduced with increasing deformation speed. Increasing the strain rate work hardens the ferrite more rapidly. This in turn displaces the deformation process towards the central part of the grains. The amount of steel participating in this process is increasing [8]. This results in a higher homogeneity of plastic deformation in CC products and limits the tendency to the transverse crack formation simultaneously [3].

The ferrite morphology influences the parameters of localised deformation realised i CC products and also the tendency to their transverse cracking. The ferrite decorating the austenite grain surfaces has the allotriomorphic feature usually, what further results in the modification of austenite phase transformation process realised in adjoining matrix regions. The physical metallurgy behaviour of the ferrite-austenite interface plays an important role as active and/or as inert interface in this connection [11]. The active allotriomorphic ferrite defined as that which is able to develop into other transformation products such as acicular Widmanstätten ferrite or bainite sheaves at the transformation temperature of interest. The allotriomorphic ferrite is said to be inert when the local reduction in transformation temperature at the ferrite interface due to the partitioning of carbon prevents the development of secondary constituents formed by displacive mechanism. The example of the inert interface is presented in Fig.1. Under this condition, the allotriomorphic ferrite can be rendered inert by the build up of carbon in the austenitic matrix ahead of the allotriomorphic ferrite-austenite boundary [11].

The corresponding controlling process is linked to the local carbon enrichment at one interface. The found asymmetric morphology represents a very interesting behaviour. The causes of this effect are not elucidated unambiguously up to this time. The following possibilities could be taken into consideration.

For example, the described response can be attributed either to the localised grain boundary enrichment of Mn what leads to a lower carbon activity or to the asymmetric cooling conditions. This interface has a "straight-line" appearance without ascertainable deviations in its course (Fig.2).

This figure shows a section of the interface enriched of carbon. On the contrary, the section without this enrichment can be described as a continuously growing acicular Widmanstätten ferrite out of the allotriomorphic intergranular ferritic zone. The feature of initiated cracks are also different. The cracks running along the "straight-line" inert interface austenite-ferrite are not accompanied with the formation of deviations (Fig.3a). These cracks have a smooth course. The cracks formed in the central part of the intergranularly precipitated allotriomorphic ferrite are accompanied with the formation of deviations in their propagation and interconnect the non-metallic inclusions (Fig.3b). In this case, the crack roughness parameter [12] is 1,25 approximately, while in the first presented example (Fig.3a) the roughness parameter of cracks is only about 1,04.

The figure 4 represents a very interesting example of the allotriomorphic ferrite initiation. The ferritic particles are formed around the non-metallic inclusions precipitated at austenite grain boundaries. In their vicinity, the traces of ferrite deformation can be observed, inclusive of small cracks initiation running from the interface of non-metallic inclusions and the ferritic matrix. The figures 5 a, b show the regions of acicular ferrite which are probably formed in deformation bands initiated in strain fields which are induced around the non-metallic inclusions. The presented microstructure examples demonstrate an immediate relation between the ferrite morphology type and the different parameters of crack formation.

Fig.5a Regions of the acicular ferrite

Fig.5b Detail of acicular ferrite initiation

4. Influence of investigated residual elements on hot ductility shortness of steels

The presented analysis is devoted to the evaluation of Cu, Sn and Sb, pertaining to the group of tramp elements, on the modification of hot tensile ductility level. The surface enrichment of these elements is a cause of their detrimental effect leading to the development of surface brittlenes in CC steel products. The surface hot shortness and the resulting increased tendency to cracking are commonly associated with the presence of liquid Cu and/or Cu-enriched phases at scale-metal interface.

The presence of tramp elements as Sn and Sb contributes to the increase of susceptibility to the surface cracking [13]. These elements reduce the Cu-solubility in steel as well as the melting temperature of formed Cu-rich phase. On the contrary, the content of Ni in steel produces the reverse effect because the addition of this element can be hold for a "remedy" ensuring the improvement of hot ductility level. The action of the mentioned elements can be exposed by means of Cu_{eq}-parameter having the following form: $Cu_{eq} = Cu + 6 \div 8(Sn+Sb)-Ni$ and/or using the relationship $Cu+6(Sn+Sb)-Ni \leq 9/E$ where E is the enrichment factor expressing the concentraction of residual elements at the interface scale-metal. The enrichment factor is defined by E = average concentration of residual elements in subscale/ bulk residual concentration [13].

Nickel increases the Cu-solubility in austenite as well as favours the occlusion of Cu in scale. This results in the modification of chemical composition of phases formed on the steel surface and having a higher melting temperature. The beneficial effect is also ascribed to the Si-influence, because this element contributes to a higher Cu-occlusion in the scale. Silicon decreases the volume fraction of Cu-rich liquid phases and simultaneously increases the level of tension necessary for intergranular penetration of considered liquid phases. On the contrary, Sn and Sb limit the extent of internal steel oxidation and owing to this effect the Cu-occlusion is decreased. In addition to this influence, Sb blocks the oxygen adsorption what reduces Cu-occlusion. However, the hot ductility tensile tests of steel, containing a higher Cu-content, performed under oxidation suppressing atmosphere (in Ar) result in attainment of higher values of RA. These tests realised in protection Ar-atmosphere do not take into account the selective oxidation of the steel surface and its subsequent influence on the hot shortness phenomena [2, 13].

The selective oxidation of steel plays a very important role as by Cu enrichment in the scale-steel interface as by the formation of the interface unevenness contributing to the development of occlusion process. Because the free enthalpy of the Cu and Ni oxides is higher than those of FeO, Cu and Ni tend to remain in the scale-steel interface without oxidation. These elements should be concentrated near the interface. It was reported that the small addition of Ni makes the scale-steel interface uneven. The same action can be ascribed to the Mn and Si action because increasing of these elements accelerates the irregularity of the scale-steel interface. It is though that besides the external oxidation the internal oxidation process also takes part in irregularity formation of the above mentioned interface [14].

The elements that are internally oxidized are Fe, Mn and Si. The internal oxides are (MnFe)O and $(MnFe)_2SiO_4$ in steel containing Si. The interface of these oxides is enriched of Ni and Cu in comparison with matrix since the diffusion coefficients of Ni and Cu in austenite are smaller than those of Fe and Mn. For this reason the concentration of Ni and Cu is increased around the internal oxides as they grow. The corresponding enrichment mechanism of Cu and Ni around the growing internal oxide of (Mn, Fe)O-type is presented in Figs.6a, b. The set of realised partial processes presented in Figs.7a, b shows the growth of internal oxides enriched of Cu and Ni. The final stage corresponds to the connection of the growing internal oxides with the scale (Fig.7c) and to the occlusion of Cu and Ni enriched particles in the scale. These process results in Cu and Ni decrease observed near the scale steel interface (Fig.7d).

Fig.6 Enrichment mechanism of Cu and Ni around the internal oxide;

a) Starting stage; b) Stage after the growth of internal oxide

The participating stages of the occlusion process are following: a) Fe is more reactive than Cu and Ni, these elements remain as metals near the scale-steel interface; b) oxygen preferentially penetrates into the area containing the enriched Ni and Cu through grain boundaries and various lattice defects. The selective oxidation is realised in grain boundaries and subsequently, the oxygen concentration increases in the grains and internal (FeMn) oxides are formed what is associated with Cu and Ni enrichment by the growth of internal oxides (Fig.6b); c) internal oxides tend to growth easily and combine with the external oxide layer. The described heterogeneous Cu and Ni enrichment results in the subsequent heterogeneous external oxidation and in the formation of uneven scale-steel interface; d) the Cu and Ni enriched particles are occluded into scale and enrichment of Cu and Ni decreases in matrix near the scale-steel interface [14]. The described occlusion process (Figs.7a-d) represents an additional mechanism to the Ni-action contributing to the increased Cu-solubility in austenite [2, 13].

5. Conclusions

The work describes the effect of microstructure parameters influencing the achieved level of hot ductility in the critical temperature region corresponding to the deformation process in heterogeneous matrix consisting of ferrite and austenite mixture. Besides this analysis, the work describes the influence of residual elements on the modification of hot ductility response. The relation between hot tensile ductility and the susceptibility to the crack formation is presented. The obtained results are preferentially analysed from view point of crack formation in cast products.

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