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Microstructural Investigations on Decomposition Behaviour of M-A Constituent and Recovery of Ductility due to PWHT in Simulated HAZ of HSLA Steels†

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Abstract

The influence of the amounts of M-A constituent in simulated HAZ on notch toughness was investigated for HSLA steels. Post-heat treatment tests at 823 and 923K were made to investigate the metallurgical behaviour and ductility change. consequently the relation between the amount of M-A constituent and recovery of notch toughness was investigated. Moreover secondary hardening and temper embrittlement appeared in steels HT100 were analyzed. Generally it can be said that, when the HAZ steels contains a large amount of M-A constituent, the improvement of notch toughness could be associated with decomposition of this phase.

KEY WORDS: Decomposition, M-A, Ductility, PWHT, HSLA.

1. Introduction

The influence of M-A constituent on mechanical properties of HAZ has been studied in some works [1-4]. The importance of the M-A constituent has been emphasized in current studies for HSLA steels [1], to find the relation between loosing toughness of HAZ in welded joints and quantity of M-A constituent in the structures. Recent investigations suggested quite clearly that M-A constituent which appeared in HAZ of weld joints of HSLA steels with medium or large weld heat input showed a deleterious effect on notch toughness. Because of the possibilities to use large heat input in welding for some constructions it is important to know conditions of recovery of notch toughness of simulated HAZ after post-heat treatment (PWHT). The investigation was carried out on four types of HSLA steels to identify the effect of PWHT on recovery of plastic properties of the HAZ structure with a considerable quantity of M-A constituent. However large attention was paid to HT100-B steel which contains some amounts of alloying elements because these elements can influence the behaviour of HAZ structure under PWHT.

2. Experimental Procedures

Four high strength low alloy steels of HT80 and experimental HT100 steels (of each two types) were used in this investigation. HT80-A (t=60 mm) and HT100 (t=150 mm) steels were subjected to each quenched and tempered treatment. HT80-B (t=25 mm) and HT100-B (t=38 mm) steels were subjected to each direct quenched and tempered treatment (DQT). The chemical compositions of each steel are presented in **Table1.**

The standard charpy-V impact specimens were used for simulating weld thermal cycles. Simulated weld thermal cycles were conducted by a dynamic thermal test machine (Gleeble-1500) which is based on direct resistance heating.

In order to investigate the relation between the continuous cooling time and formation behaviour of M-A constituent for the purpose of improvement on notch toughness, two types of thermal cycles are used in the investigation. The thermal cycles are shown in **Fig. 1 (a)** and **(b).** Charpy impact test was used to evaluate the notch toughness of the specimens as absorbed energy at room temperature by a 480J tester in maximum.

The hardness was evaluated by Vickers micro-hardness tester. TEM (thin films and carbon replicas) was used to analyze microstructure changes.

3. Experimental Results

3.1 Ductility investigation

The influence of M-A constituent on impact values in all

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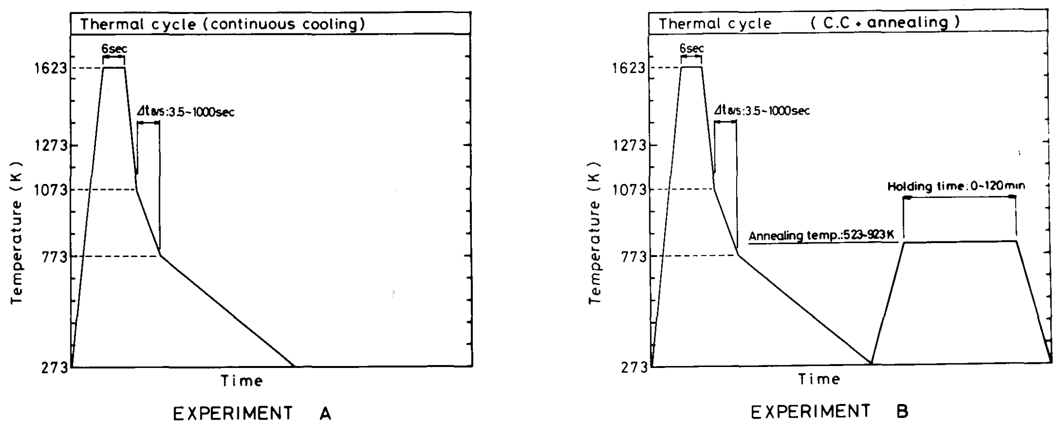
Table 1 Chemical compositions of steel plates used

Material	Thickness mm	Chemical compositioins (wt%)													Ceq	Pcm
		C	Si	Mn	P	S	Cu	Ni	Cr	Mo	V	Nb	Al	B		
HT80-A	60	0.12	0.23	0.93	0.006	0.002	0.19	1.22	0.48	0.45	0.03	–	0.066	0.0007	0.53	0.26
HT80-B	25	0.10	0.27	0.92	0.009	0.002	0.20	0.98	0.47	0.27	0.04	–	0.052	0.0012	0.45	0.23
HT100-A	150	0.14	0.05	1.03	0.004	0.001	0.23	3.80	0.56	0.58	0.027	0.012	0.047	0.0011	0.67	0.34
HT100-B	38	0.11	0.11	0.86	0.004	0.001	0.03	1.57	0.54	0.56	0.073	–	0.053	0.0011	0.55	0.26

HT80-A, HT100-A: Quench and Tempered
HT80-B, HT100-B: Direct Quench and Tempered



(a) continuous cooling method (b) PWHT method after continuous cooling

Fig. 1 Simulated thermal cycle
(a) continous cooling method
(b) PWHT method after continuous cooling

mentioned steels with simulated HAZ was studied in the previous work [1]. It was found out that M-A constituent appeared in higher heat input than about 100s in Δt8/5. The amount of M-A constituent was increasing with longer cooling time and accordingly the impact values were reduced.

In this study a thermal cycle of Tp = 1623K and Δt8/5 = 200s was chosen to investigate the influence of PWHT on ductility of HAZ. It was found out that the content of M-A constituent at Δt8/5 = 200s is large enough to decrease notch toughness in maximum way as shown in **Fig. 2**[**1**]. Two temperatures were chosen, 823 and 923K, for PWHT to improve the ductility of simulated HAZ. The time for PWHT was selected from 0.5 to 16 hours. The changes in notch toughness in dependence on time and temperature of all steels are shown by solid lines on **Fig. 3** (**a**) to (**d**). The changes in hardness are marked by broken lines on individual Fig. 3 (a) to (b).

The influence of the heat treatment on ductility can be seen clearly. Direct improvement of notch toughness is on steels HT80-A and HT80-B what is in good correlation with changes of hardness. Decreasing in ductility of steel HT100-A after heat treatment at 823K is evident although there is decreasing in hardness like the results of steels HT80-B and HT80-A. More interesting changes are on the steel HT100-B. Ductility was improved at the beginning of heat treatment (0.5 and 1 hour). Later there is a drop in notch toughness at 923K between 2 and 4 hours followed again by improvement. After improvement of notch toughness at beginning at 823K there is a drop in ductility over the whole time of the heat treatment. Quite important fact is that the hardness is not decreased so continuously at 923K and nearly no decrease in hardness at 823K as on the other steels.

3.2 Microstructure and fracture analyses

Generally optical microstructure of the simulated HAZ of HT100-B is characterized on **Fig. 4**. The structure of

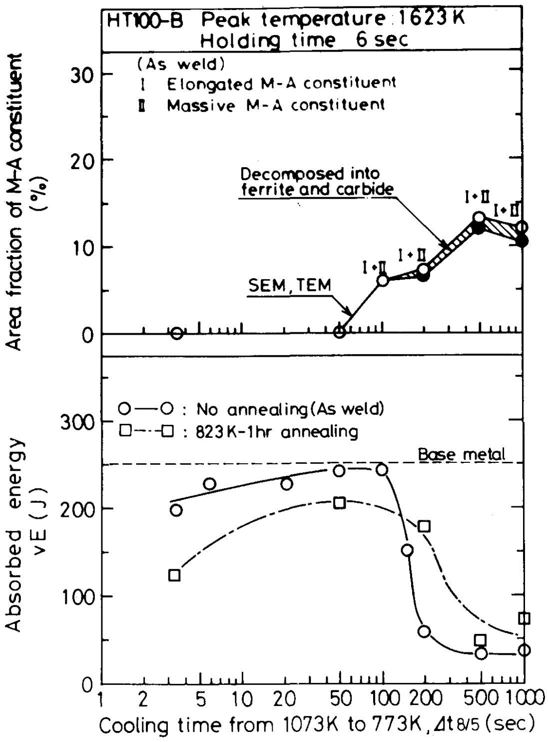


Fig. 2 Area fraction of M-A and absorbed energy as function of continuous cooling time

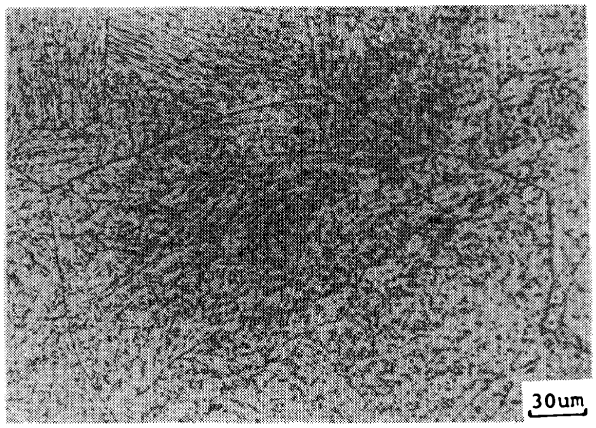


Fig. 4 Microstructure of HT100-B steel after the simulation of thermocycle Tmax = 1623K, Δt8/5 = 200s

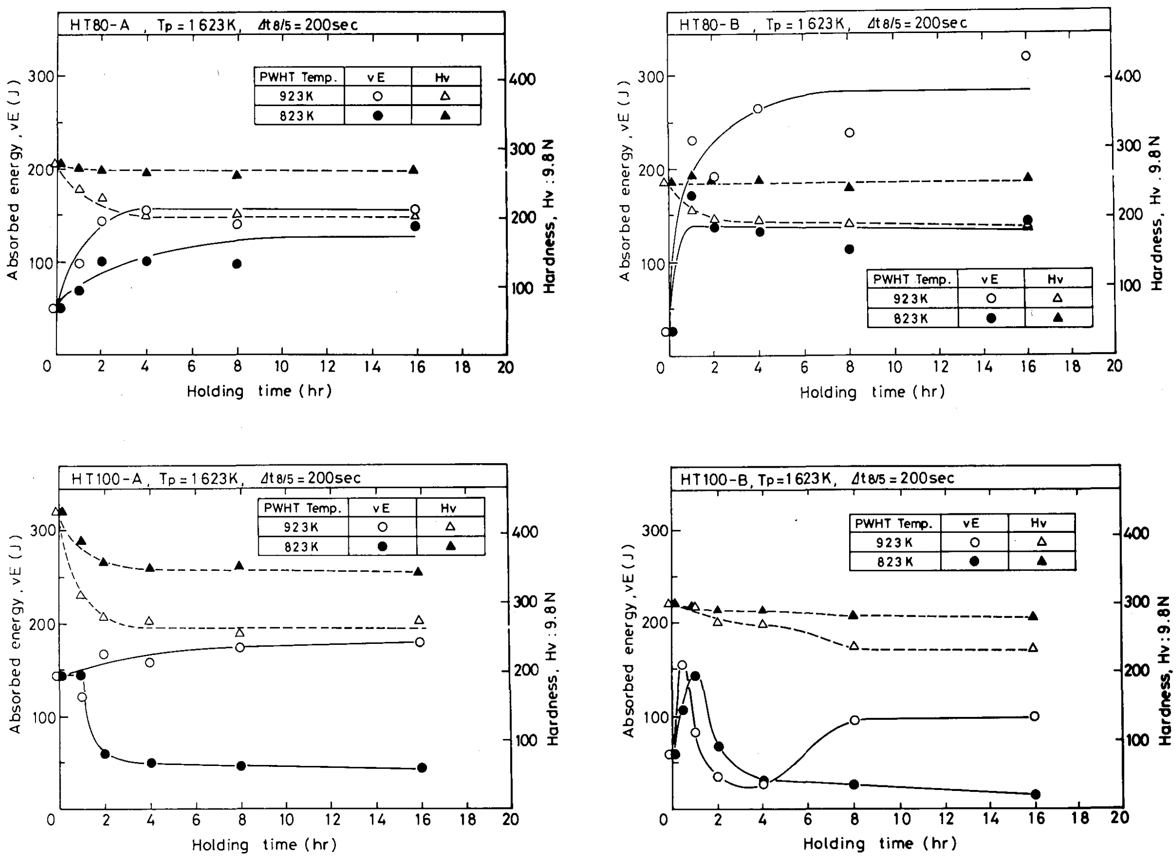


Fig. 3 The changes of absorbed energy and hardness during post heat treatment

simulated HAZ of steels consists of upper bainite I and size of grains is ≒200 μm. Microstructure of steel HT100-A consists of lower bainite and grain size is ≒200 μm. The influence of heat treatment on TEM microstructure at temperatures 923K for HT100-B in dependence on time is characterized on **Fig. 5.** It is clearly seen the progress of the decomposition of M-A constituent with time. The time-temperature influence on the decomposition of M-A constituent was seen to be agreed with the change on **Fig. 6** (a) to (c). These microstructural changes correlate well with decreasing in hardness and improving of notch toughness of HT80-A and B steels. The structure of HAZ in HT100-A steel consists of lower bainite without M-A and of course there are not so large microstructure changes but only some spherodizing of cementite in dependence on heat treatment. These changes are more in progress at 923K than at 823K. This is confirmed by measurement of hardness too. However these results are in disagreement with results of impact test after heat treatment at 823K as shown in Fig. 3 (c).

More complex behaving was showed by impact test of HT100-B steel. It seems to be like secondary hardening in consideration of measured results of hardness (Fig. 3 (d)). Microstructure was analyzed after heat treatment at 923K-8hr and 16hr where is clear improvement of impact values and decreasing in hardness. Carbon extraction replicas and thin foils technique were used to control possible presence

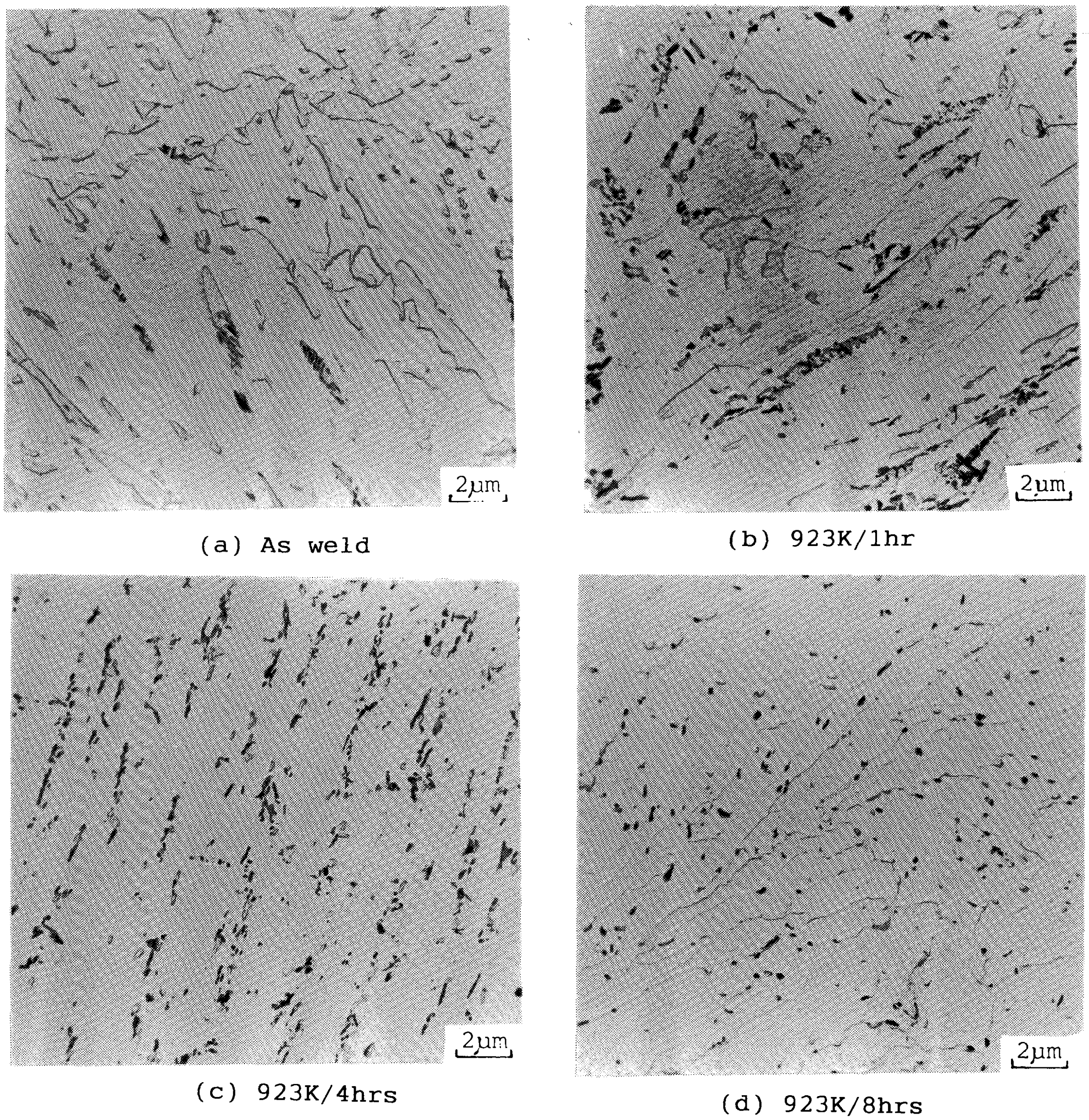
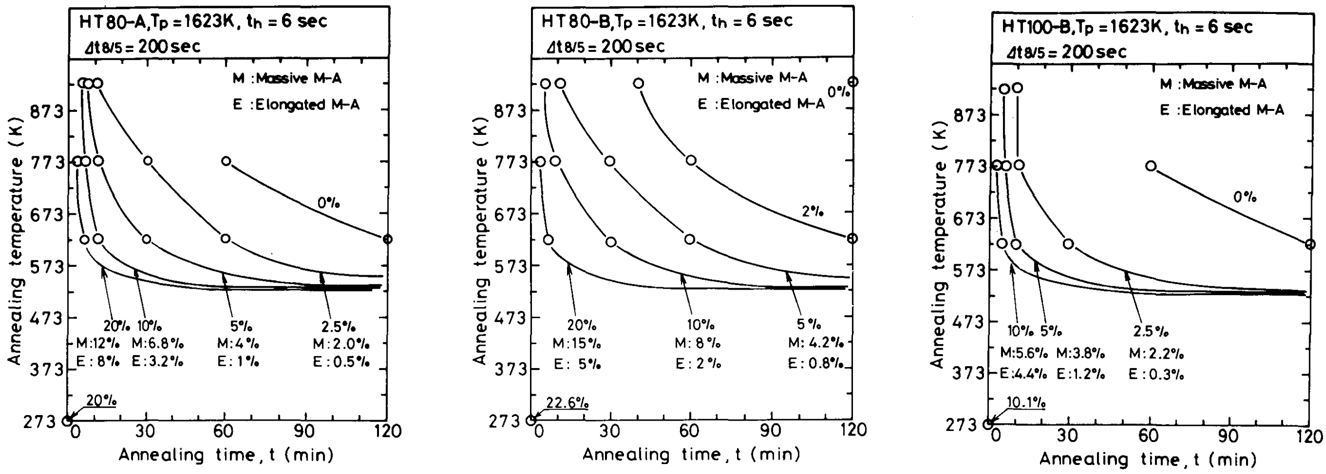


Fig. 5 Microstructural changes after heat treatment at 923K



(a) HT80-A (b) HT80-B (c) HT100-B

Fig. 6 Decomposition behaviour of M-A constituent as a function of PWHT time and temperature

of fine carbides which can be responsible for the slight secondary hardening because of low content of vanadium. Fine non-coherent MC carbides were analyzed after heat treatment at 923K-8hr (**Fig. 7**). The microstructural analysis of MC carbides was made on carbon extraction replica. The carbides consists of V, Cr, Fe and some traces of Mo as shown in **Fig. 8**.

Fracture surfaces of impact specimens were investigated over all steels. Generally it can be said that the quantity of ductile dimple fracture has been increased with improving of impact values. However a little differences were seen at steel HT100-B. There was higher amount of dimple fracture after short time of heat treatment (0.5, 1hr) what is in a good correlation with the impact values. After prolongation of heat treatment the dimple fracture was reduced that the initiation area of the fracture at 923K-4hr was cleavage and at 823K-4hr cleavage and intergranular as shown in **Fig. 9**. The intergranular facets have been seen only in initiation area of the fracture surfaces as shown in Fig. 9. No intergranular facets were observed in other parts of the fracture even at longer time.

Fracture surfaces of impact specimens of HT100-A after heat treatment at 823K has been investigated too and intergranular facets were found out as shown in **Fig. 10**. The quantity of the intergranular facets was increased with increasing in time. The Auger spectroscopy analysis was made on intergranular fracture surfaces after heat treatment at 823K-24hr. The results are on **Fig. 11**. The presence of P and B can be seen clearly.

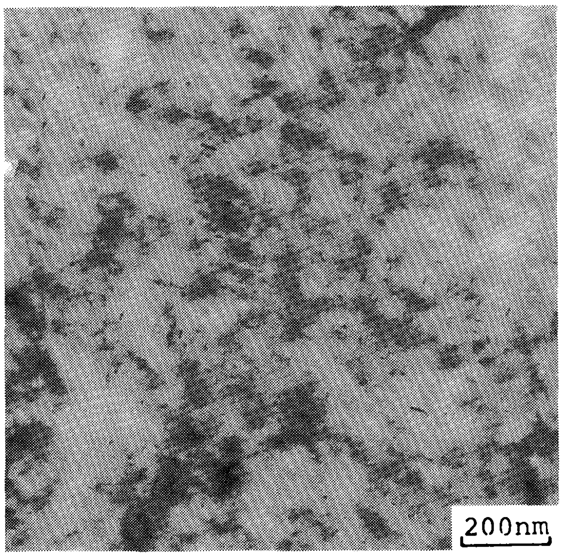


Fig. 7 MC carbides in matrix of HT100-B after heat treatment 923K/8hr

4. Discussion

As it was said at beginning the M-A constituent which is

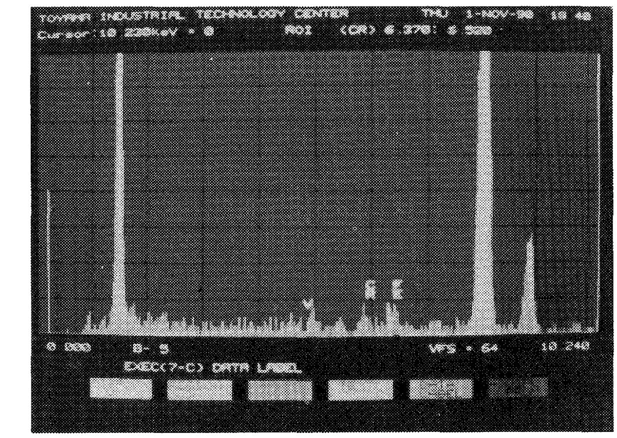


Fig. 8 Microchemical analysis of MC carbides extracted on carbon replica

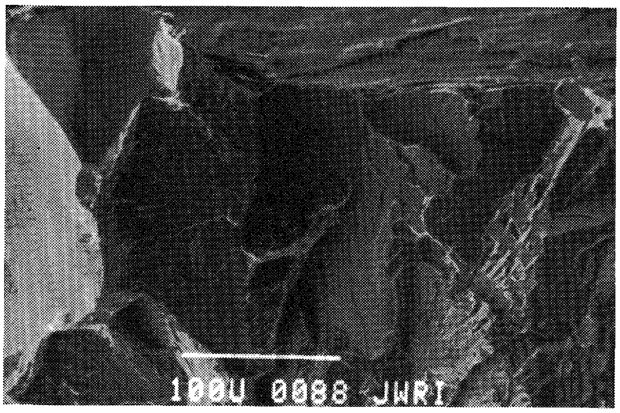


Fig. 9 Fracture surface of the impact specimen of the steel HT100-B after heat treatment 823K/4hr

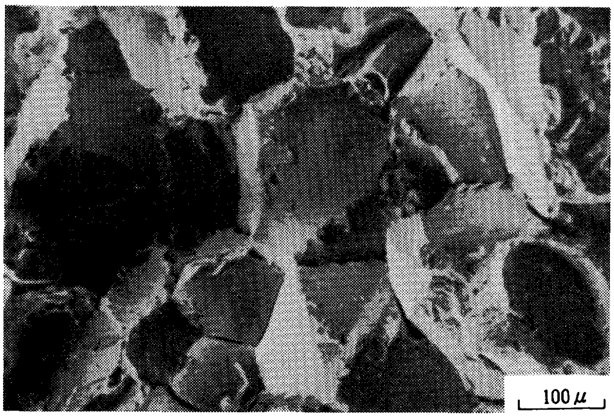


Fig. 10 Fracture surface of the impact specimen of the steel HT100-A after heat treatment 823K/8hr

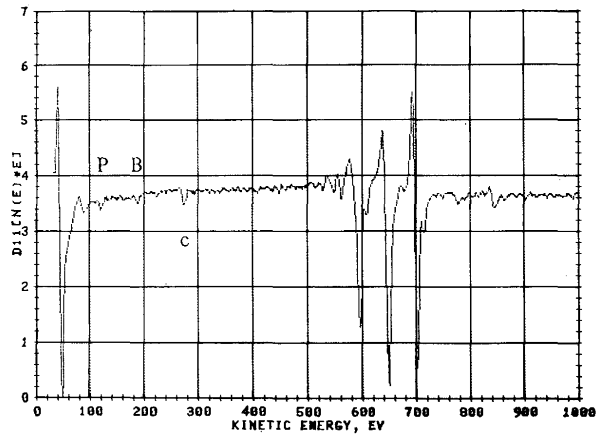


Fig. 11 Result of Auger spectroscopy analysis on intergranular fracture surface of the steel HT100-A after heat treatment at 823K/24hr

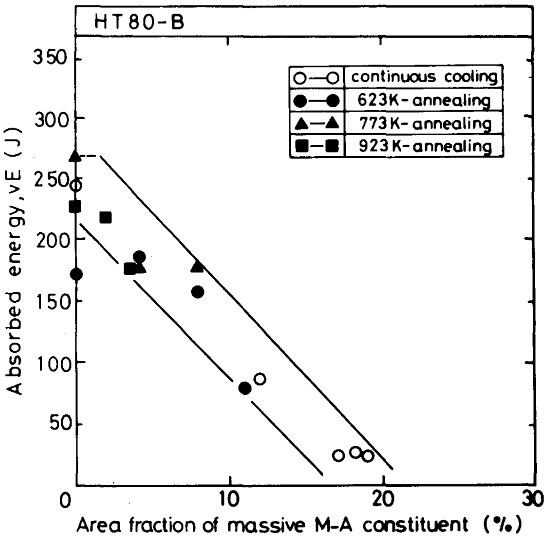


Fig. 12 Relation between absorbed energy and area fraction of massive M-A constituent

a part of upper bainite structure has detrimental effect on the plastic properties of HAZ. This influence can be explained on the basis of the Ashby model of deformation of non-homogenous materials [5]. Generally it can be said, when the matrix is deformed in which second phase is contained, that the particles of M-A constituent will work as obstacles for moving of dislocations. Depending on size and quantity of M-A constituent there is a large inhomogeneity in plastic deformation. Then when the critical density of dislocations is reached in surrounding M-A constituent. The micro-cracks can appear very quickly. The influence of this inhomogeneity is that the micro-crack can propagate quite easy. These could be the reason of such detrimental effect of M-A constituent on notch toughness particularly when the grain size is so large as in this case ($>200\,\mu$m). Microstructural conditions are expressively changed under the heat treatment because of transformation of M-A constituent to cementite carbides and ferrite. That means instead of one piece of M-A constituent there are many finer carbides (Fig. 5). Under the Ashby model [5] the plastic deformation should be spread more homogeneously and their notch toughness is increased. This effect is well seen from Fig. 3. The absorbed energy has a clear linear relation to area fraction of massive M-A constituent for both cases of continuous cooling and for PWHT conditions as shown in **Fig. 12** as for HT8-B.

Influence of this phenomena can be seen at the beginning of the heat treatment of HT100-B steel. Next decreasing in impact values is influenced by secondary hardening of MC carbides where M consists of V, Cr, Fe and traces of Mo. The question is why precipitation of MC carbides needs so long time at 923K. The answer could be in the distribution of carbon and alloying elements. It is clear that carbon is concentrated in M-A constituent [1] and alloying elements are spread homogeneously in matrix in cooling time of $\Delta t8/5=200$s. At the beginning of heat treatment M-A constituent was changed to cementite carbides and ferrite. Although vanadium is strong carbide former element it needs time to form MC carbides because of low content of V and the free quantity of carbon in solid solution in matrix was low too. The speed of secondary hardening is clearly seen from hardness-temperature dependence. This process is longer for 823K. When it is used Hollomon-Jaffe parameter P the time for heat treatment at 823K will be calculated from dehardening area of heat treatment at 923K-8hr. It will be concluded that the time of hundreds hours is needed at 823K. That is an explanation of very low impact values at longer time at 823K. The presence of some intergranular facets (HT100-B) in the initiation area (Fig. 9) could be explained by influence of the grain size and the role of grain boundary at the beginning of plastic deformation in hardened matrix [7].

Loosing of notch toughness of HT100-A is clearly explained by temper embrittlement which was influenced by segregation of elements P and B. Fig. 11 shown an example of the Auger spectrographs of intergranular fracture surface which was obtained on the simulated HAZ of HT100-A specimen after heat treatment 823K-24hr. Segregation of P and B was clearly observed.

5. Conclusions

The weld thermal cycle simulation and post-heat treatment tests were made to investigate the relation between the behaviour of formation and decomposition of M-A constituent.

Main results obtained are as follows;

(1) Direct dependence between the quantity of M-A constituent in ferrite of simulated HAZ of steels HT80-A, B and impact values was found out.

(2) Heat treatment at 823K and 923K was used to improve plasticity of simulated HAZ. Notch toughness was increased in direct relation with decomposition of M-A constituent (steels HT80-A and B).

(3) The effect of secondary hardening by MC carbides appeared at steel HT100-B during heat treatment although the content of V is low (0.013%).

(4) Temper embrittlement appeared at steel HT100-A during the heat treatment on 823K. The reason of this temper embrittlement was segregation of P and B on grain boundaries.

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