EFFECT OF PURITY AND THERMOMECHANICAL TREATMENT ON FRACTURE BEHAVIOUR OF LOW-ALLOY MARTENSITIC STEELS

J. Eickemeyer and G. Zouhar

Akademie der Wissenschaften der DDR, Zentralinstitut für Festkörperphysik und Werkstofforschung, Dresden, German Democratic Republic

ABSTRACT

Strength, fracture toughness and stress corrosion cracking threshold of tempered martensitic silicon-manganese steels were analysed in relation to local failure mechanisms after high temperature thermomechanical treatment (HTMT) and conventional treatment (CT). In order to investigate the role of minor impurities two material qualities containing different amounts of minor impurities are compared. The results indicate that the fracture toughness K_{IC} at constant strength level is markedly enhanced by the HTMT applied and also by decreasing the level of minor impurities. In comparison with these effects on K_{IC} the influence of HTMT on the stress corrosion cracking threshold K_{ISCC} is far less pronounced. An influence of the decreased impurity content on K_{ISCC} did not appear. It is concluded that the HTMT suppresses the enrichment of detrimental impurities at the austenite grain boundaries with the consequence of transcrystalline plastic fracture at unstable crack growth in the materials of commercial purity. However, this purification of grain boundaries by HTMT produces no marked effect on the stress corrosion cracking behaviour because of the cooperative action of hydrogen and detrimental impurities, which yield intercrystalline brittle fracture.

KEYWORDS

Martensitic steels; minor impurities; segregations; conventional treatment; thermomechanical treatment; fracture toughness; stress corrosion cracking; hydrogen embrittlement.

INTRODUCTION

Primary recrystallization of prior austenite in low tempered low alloy martensitic steels during HTMT may yield intercrystalline brittle fracture during unstable crack growth with relatively low fracture toughness K_{IC}. This was observed on a 0,5%C-1,5%Si-0,7%Mn- and a 0,5%C-1%Cr-0,1%V-steel (Zouhar jr., 1979, 1981). It was suggested, that critical enrichments of detrimental impurities at the austenite grain boundaries appear by nonequilibrium segregation during primary recrystallization (Wieting, 1976, 1979; Beyer, 1981). If this enrichment is transmitted to the martensite during quenching it decreases the cohesion of the prior austenite grain boundaries. Furthermore from the results on the 0,5%C-1%Cr-0,1%V-steel it was concluded, that lattice defects introduced in the stable austenite by comparatively small amounts of plastic deformation during the last pass at HTMT without initiation of primary recrystallization increase the matrix solubility, and some purification of the grain boundary results. This is connected with transcrystalline plastic fracture and results in an increased fracture toughness, K_{IC}, as well as an increased stress corrosion cracking threshold, K_{ISCC}, (Zouhar jr. and co-workers, 1981). Investigations as to what extent this purification mechanism by HTMT may be applied also to low alloy Si-Mn-steels are not known. In this paper results about the influence of HTMT and CT on strength, toughness and stress corrosion cracking behaviour of high strength martensitic Si-Mn-steels are presented. In order to elucidate effects of minor impurities on the fracture toughness and the stress corrosion cracking threshold two steel qualities with different purity levels are examined.

EXPERIMENTS

The chemical composition of the Si-Mn-steels used in this investigation is shown in Table 1. The abbreviation CP stands for the arc-melted material of commercial purity.

TABLE 1 Chemical Composition of the Si-Mn-Steels (m-%)

No.	Material	C	Si	Mn	P	S	Cr	Cu	Ni
1	65SiMn7 (CP)	0,59	1,62	0,73	0,014	0,025	0,18	-	-
2	65SiMn5 (CP)	0,56	1,10	1,19	0,021	0,036	0,21	-	-
3	50SiMn7 (CP)	0,47	1,84	0,68	0,020	0,034	0,12	-	-
4	" "	0,50	1,54	0,68	0,026	0,040	0,13	0,17	0,08
5	" (HP)	0,42	1,44	0,71	0,005	0,006	0,03	0,02	0,04

- not determined

The HP-quality was melted in a 10kg-vacuum induction furnace from high-purity base materials.
Rolled billets of 38 mm diameter (materials No. 1,2,3) were rolled down on a HTMT-experimental rolling mill (Schmitt and co-workers, 1976) in thirteen passes (austenitizing temperature ≈ 1250°C, finish-rolling temperature ≈ 850°C) to rods with an oval sectional area of about 50 mm² for prestressed concrete. The reduction of cross sectional area in the last two passes amounted to about 10%. Immediately after rolling the material was quenched in water, followed by an inductive shock tempering treatment at different temperatures between 450 to 615°C (time of tempering about 2 s) to get different strength levels. Further details on the deformation-temperature-time-regime of HTMT applied to materials No. 1, 2 and 3 are given elsewhere (Zouhar and co-workers, 1983). The conventional heat treatment was carried out by austenitizing at about 850°C, followed by oil quenching and bath annealing at 300, 400 and 480°C to obtain different strength levels. Details of the HTMT- and CT-regimes applied to materials No. 4 and 5 are described in another paper (Zouhar jr. and co-workers, 1979). The HTMT and CT were carried out on flat rods. These rods were tempered at 200°C, for 1 hr, after quenching from finish-rolling temperature (HTMT) and austenitizing temperature (CT).

The specimens for tensile tests were prepared with their axes parallel to the rolling direction. The orientation of the fatigue precrack for the fracture mechanics SEN bending specimens was at right angles both to rolling direction and plane. The threshold for stress corrosion cracking K_{ISCC} was measured in a saturated $Ca(OH)_2$ + $CaSO_4$-solution (p_H = 12,5). Details of the test method are published elsewhere (Eickemeyer and co-workers, 1974). All tests were conducted at room temperature.

RESULTS AND DISCUSSION

Figure 1 illustrates the fracture toughness-yield stress and the stress corrosion cracking threshold-yield stress relations for the CP-materials 50SiMn7, 65SiMn5 and 65SiMn7, resulting from HTMT and CT.

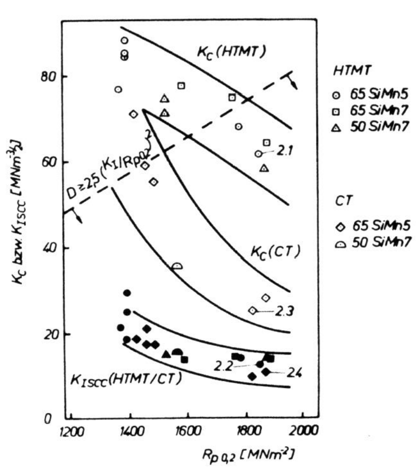

Fig. 1. Fracture toughness K_C and stress corrosion cracking threshold K_{ISCC} dependence on the yield stress $R_{p0,2}$ for different Si-Mn-steels after HTMT and CT. Numbers 2.1. to 2.4 are related to scanning electron micrographs in Fig. 2.

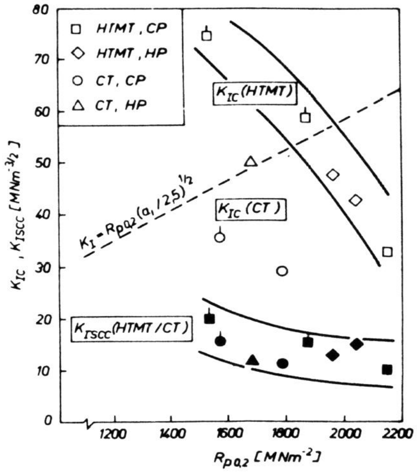

Fig. 3. Fracture toughness K_{IC} and stress corrosion cracking threshold K_{ISCC} dependence on the yield stress $R_{p0,2}$ for 50SiMn7-steel of different purity after HTMT and CT. Dashed symbols from Fig. 1. HP - high purity CP - commercial purity.

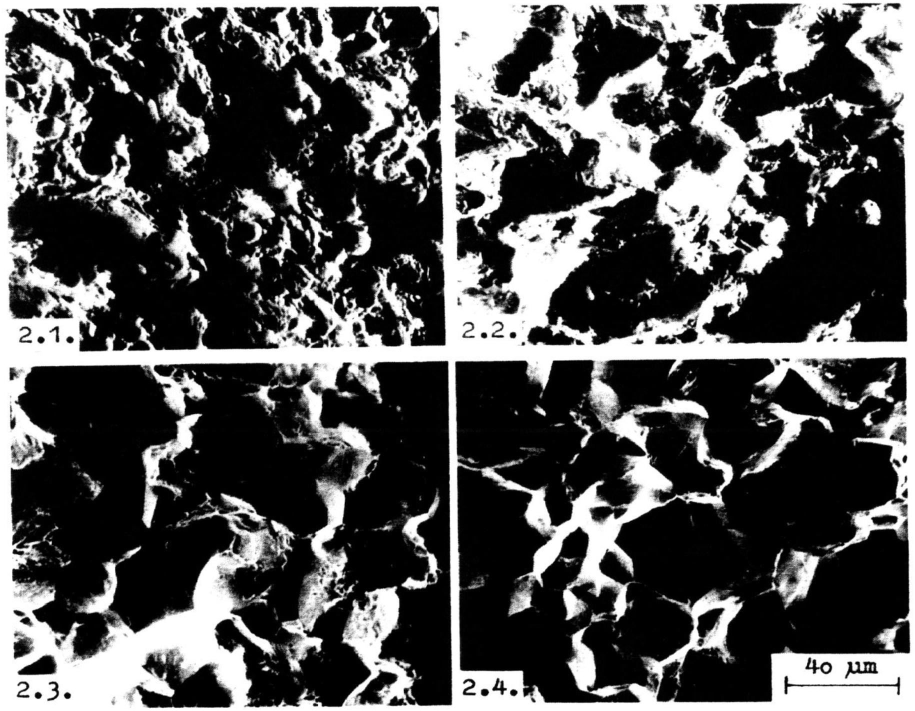

Fig. 2. Scanning electron micrographs of 65SiMn5-steel after K_{IC}- and K_{ISCC}- testing. 2.1.- K_{IC}/HTMT; 2.2.- K_{ISCC}/HTMT; 2.3.- K_{IC}/CT; 2.4.- K_{ISCC}/CT. Numbers are related to Fig. 1.

If the yield stress $R_{p0,2}$ is increased by 35% with reference to $R_{p0,2}$ = = 1400 MNm^{-2} the fracture toughness is lowered by nearly 20% after HTMT, however by about 60% after CT. The HTMT applied enhances the fracture toughness by 100% with respect to the fracture toughness after CT at the same yield stress in the range of 1600 to 1900 MNm^{-2}. This effect is connected with a change from intercrystalline brittle fracture after CT to transcrystalline plastic fracture after HTMT. Figures 2.1. and 2.3. give an example of this change in fracture mechanism during unstable crack growth for the material 65SiMn5 at constant yield stress $R_{p0,2}$ = 1900 MNm^{-2}. At a lower yield stress $R_{p0,2}$ = 1450 MNm^{-2} HTMT has no effect on fracture toughness with respect to CT. In this case transcrystalline plastic fracture appeared after HTMT as well as CT.

Furthermore, Fig. 1 shows that the HTMT also affects the stress corrosion cracking threshold K_{ISCC}. However, this effect is observed only at high yield stresses and is very low in comparison with the influence of HTMT on the fracture toughness. The K_{ISCC}-values are enhanced by about 30% in consequence of HTMT compared to the K_{ISCC}-values after CT at yield stresses $R_{p0,2}$ = 1800 to 1900 MNm^{-2}. This result corresponds with the different fracture behaviour after HTMT and CT during subcritical crack growth. Whereas a mixed fracture mode prevails after HTMT (Fig. 2.2.), a pronounced intercrystalline brittle fracture with very small plastic areas predominates after CT (Fig. 2.4.).

Figure 3 shows the fracture toughness K_{IC} and stress corrosion cracking threshold K_{ISCC} in relation to the yield stress $R_{p0,2}$ for the steel 50SiMn7 with high and commercial purity after HTMT and CT. After CT the HP- material indicates an increased fracture toughness in comparison with the K_{IC}- values of the CP- material. The HP- material shows transcrystalline plastic fracture. The fracture toughness- yield stress relations of the HP- material after HTMT fit in the corresponding K_{IC}- $R_{p0,2}$- dependence of the CP- materials (Fig. 1.). Independent of HTMT and CT an influence of the decreased impurity content on the stress corrosion cracking threshold is not observed.

Above results reveal that the HTMT controls the cohesion of the prior austenite grain boundaries in high strength martensitic low-alloy Si-Mn-steels. Grain refinement and/or polygonized dislocation substructures developed in stable austenite during HTMT control the kinetics of the martensitic transformation and the accommodation processes during the transformation in such a way that results a fine dispersed martensite with homogeneously distributed and decreased local stresses and strains (Winderlich and co-workers, 1982; Zouhar and co-workers, 1983; Zouhar, 1984). These factors are not very likely as reasons for the observed increase in fracture toughness and stress corrosion cracking threshold by the used HTMT. Polygonized dislocation substructures are not expected in the hot rolled condition of the steel for prestressed concrete (Zouhar and co-workers, 1983), and nearly the same prior austenite grain size is observed after HTMT and CT. Therefore it has to be suggested that the different fracture behaviour of the Si-Mn- steels with commercial purity after HTMT and CT is caused by different enrichments of detrimental minor impurities at the grain boundaries in the stable austenite.

It is well known that critical enrichments of minor impurities may be produced at austenite grain boundaries by equilibrium and nonequilibrium segregation. Figure 4 shows these mechanisms schematically. If such critical enrichments are transmitted to the martensite during quenching they decrease the cohesion of the prior austenite grain boundaries. The comparatively low temperature of austenitization at CT supports the development of critical enrichments at the austenite grain boundaries of the CP- materials by equilibrium segregation with the consequence of intercrystalline brittle fracture at unstable crack growth (Fig. 2.3.).

It is expected, that for conditions examined grain boundary segregations of substitutional elements like phosphorus (Erhart and co-workers, 1983) and their interaction with copper (Wieting, 1979) are of concern. Sulphur should be bound to manganese. By decreasing the content of minor impurities the K_{IC}- values are increased for the CT condition (Fig. 3, HP- material) and transcrystalline plastic fracture is observed.
With respect to the HTMT it is assumed that the recrystallization is influenced by the applied rolling procedure in such a way that the enrichment by nonequilibrium segregation is suppressed and the grain boundary weakening is avoided. Lattice defects introduced by the comparatively low level of plastic deformation applied during the last passes are certainly not efficient with respect to recrystallization, but increase the matrix solubility (Pichard and co-workers, 1975; Wieting, 1979), and some purification of the grain boundaries results along with transcrystalline plastic fracture, (Fig. 2.1.), and markedly enhanced fracture toughness (Fig. 1.). This purification mechanism by HTMT appears as a beneficial

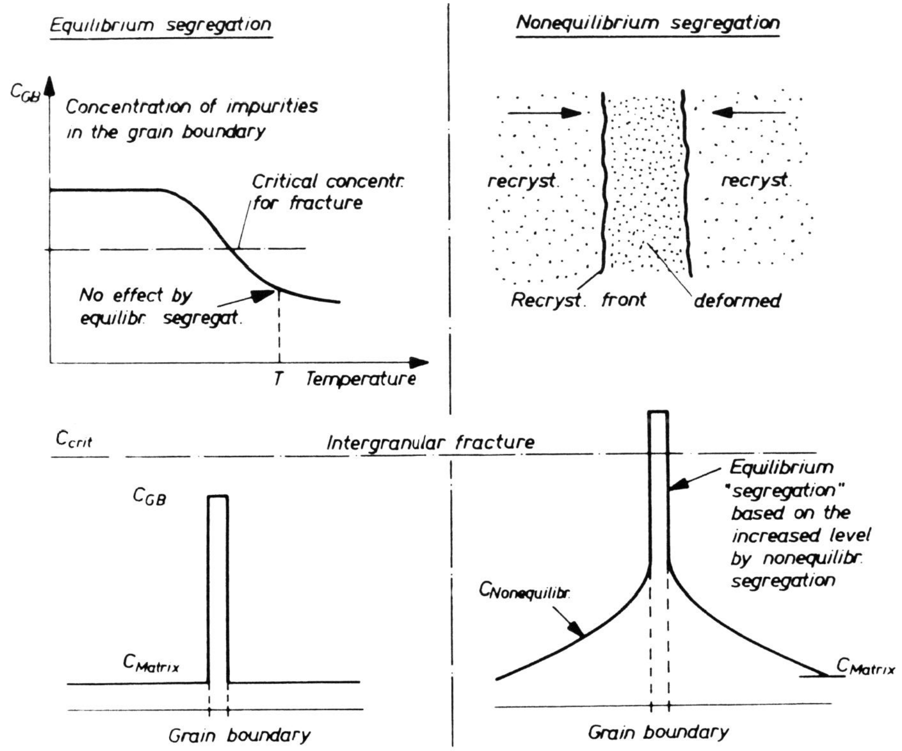

Fig. 4. Enrichment of minor impurities at austenite grain boundaries by equilibrium segregation (left part) and by nonequilibrium segregation during primary recrystallization (schematically).

way to control detrimental effects of minor impurities on the fracture resistance of high strength low alloy Si-Mn-steels during unstable crack growth.

From the absence of a purity effect on the stress corrosion cracking threshold, K_{ISCC}, it is concluded that a reduction of the phosphorus and sulphur content to 50...60 Mppm is not sufficiently to influence K_{ISCC}. By variation of the phosphorus and sulphur contents, respectively, between 20 and 300 Mppm in an AISI 4340 steel an influence on K_{ISCC} also did not appear (Sandož, 1971). In comparison with phosphorus and sulphur the alloying elements manganese and silicon should have an overwhelning effect on K_{ISCC} (Banerji and co-workers, 1978). These elements segregate also, whereby a comparatively strong interaction between manganese and phosphorus has to be taken into account (Guttmann, 1976). In addition silicon increases the solubility of hydrogen in steel (Szczepanski, 1963). In this way manganese and silicon may support the hydrogen induced fracture by decreasing the critical contents of minor impurities to lower levels with respect to K_{ISCC}. This was proved by removal of manganese and silicon from a low alloy- steel (Banerji,

1978). Thereby the stress corrosion cracking threshold K_{ISCC} was increased five- to sixfold, although the steel contained 30 Mppm phosphorus and 30 Mppm sulphur. Therefore an increase of the K_{ISCC}-values of the used Si-Mn- steels probably requires such decreased contents of minor impurities, which are scarcely of interest on technical conditions.

It is suggested, that the small increase of the K_{ISCC}- values by HTMT is caused by the distribution of minor impurities and by a small anisotropy effect, such that longitudinal cracks parallel to rolling plane appear after K_{ISCC}- testing.

CONCLUSIONS

1. The fracture toughness of low-alloy martensitic Si-Mn-steels may be increased by about 100% at nearly the same strength level by HTMT with respect to fracture toughness after CT. A reduction of the level of minor impurities also enhances the K_{IC}- values.

2. In comparison with these effects on K_{IC} the influence of HTMT on the stress corrosion cracking threshold is far less pronounced. An influence of the decreased impurity content on K_{ISCC} did not appear.

3. It is concluded that the HTMT applied suppresses the enrichment of detrimental minor impurities at the austenite grain boundaries with the consequence of transcrystalline plastic fracture during unstable crack growth in the Si-Mn- steels with commercial purity. This purification mechanism by HTMT appears as a beneficial way to control detrimental effects of minor impurities on the fracture resistance of high strength low alloy Si-Mn- steels during unstable crack growth.

4. However, this purification of grain boundaries by HTMT produces no marked effect on the condition of stress corrosion cracking because of the cooperative action of hydrogen and detrimental impurities, which yield intercrystalline brittle fracture. Manganese and silicon should enhance the sensitivity of this type of steel to hydrogen induced fracture by decreasing the critical contents of minor impurities to very low levels with respect to the initiation of stable crack growth (K_{ISCC}). Therefore an increase of the K_{ISCC}- values of the Si-Mn- steels probably requires such low contents of minor impurities, which are scarcely of interest on technical conditions.

REFERENCES

Banerji, S.K., C.J. McMahon, Jr., and H.C. Feng (1978). Intergranular Fracture in 4340-Type Steels: Effect of Impurities and Hydrogen. Metallurg. Transact., 9A, 237-247.

Beyer, B. (1980). Zur Wechselwirkung von Fremdatomen mit Korngrenzen. Dissertat., Akademie der Wissenschaften der DDR.

Eickemeyer, J., und F. Schlüt (1974). Eine Methode zur Spannungsrißkorrosionsprüfung auf der Grundlage der Bruchmechanik. Neue Hütte, 19, 232-236.

Erhart, H., and H. Paju (1983). Phosphorus segregation in austenite. Scripta Met., 17, 171-174.

Guttmann, M. (1976). Metal Science, 10, 337-341.

Pichard, C., M. Guttmann, J. Rieu, and C. Goux (1975). Les Joints intergranulaires dans les Metaux. Journal de Physique, 36, Colloque C-4 supplement no 10, 151-155.
Sandoz, G. (1971). Metallurg. Transact., 2, 1055-1063.
Schmitt, W., und G. Zouhar jr. (1976). Anwendung der hochtemperaturthermomechanischen Behandlung an Walzstraßen. Neue Hütte, 21, 193-197.
Szczepanski, M. (1963). The Brittleness of Steel. John Wiley and Sons, Inc., 172.
Wieting, J., B. Beyer, E.-H. Herbst (1976). Proceed. of the 4th Internat. Conf. on the Strength of Metals and Alloys, Vol. 3, 1280, Nancy.
Wieting, J. (1979). Einfluß von Spuren- und Begleitelementen. In: Wissenschaftliche Berichte der Akademie der Wissenschaften der DDR, ZFW Dresden (Ed.), 13. Metalltagung in der DDR, Mechanisches Verhalten von Eisenwerkstoffen, 295-306.
Winderlich, B., und G. Zouhar (1982). Zur Kaltumformbarkeit niedrigangelassener hochtemperaturthermomechanisch behandelter martensitischer Federstähle. Neue Hütte, 27, 265-270.
Zouhar Jr., G., P. Finke, A. Güth, M. Schaper and H.-J. Klauß (1979). Relation between Structure and Mechanical Properties of High-strength Low-alloy Martensitic Steels after High Temperature Thermomechanical Treatment. Proceed. of the 5th Internat. Conf. on the Strength of Metals and Alloys, Vol. 2, 1377-1382, Aachen.
Zouhar Jr., G., M. Schaper, J. Eickemeyer and P. Finke (1981). Microstructure and Fracture Mechanisms in a High Strength Low-Alloy Martensitic Steel. Proceed. of the 5th Internat. Conf. on Fracture, Vol. 2, 747-754, Cannes.
Zouhar, G., H. Wadewitz und A. Güth (1983). Martensitumwandlung - ein Indikator für die Warmformgebung. Neue Hütte, 28, 292-298.
Zouhar, G. (1984). Niedrigangelassener hochfester Martensit - ein Anwendungsgebiet für die HTMB. Freiberger Forschungshefte (demnächst).