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Mechanical properties and strain-induced phase transformations of some high-strength manganese steels

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Work has been carried out to determine the influence of composition and heat treatment on the mechanical properties of some high-strength manganese steels. These steels display impressive mechanical properties in the as-hot-rolled and solution-treated conditions. These impressive mechanical properties have been directly related to the strain-induced phase transformation which was monitored continuously using a magnetic reluctance technique *in situ* and dynamically during the mechanical tests. Exceptional work-hardening rates were displayed by these steels exceeding that of Hadfield's manganese steel, which is itself known as a highly work-hardening material. They underwent considerable phase transformation during deformation. The results indicated that work-hardening effects probably arise from an increased dislocation density and stacking fault population which have been generated in the austenite/epsilon matrix during the strain-induced formation of lath martensite which itself also takes part in the work-hardening process.

1. Introduction

Many studies have been devoted to the mechanical properties of alloys which exhibit strain-induced phase transformations. A strain-induced phase transformation has been shown to improve tensile strength and ductility of a class of steels commonly referred to as "TRIP" steels, e.g. Zackay *et al.* [1]. The mechanical properties encountered in "TRIP" steels could be obtained in some high manganese steels which had unstable austenite/epsilon martensite microstructures and which underwent strain-induced phase transformations. Unstable austenite/epsilon steels have the potential for transforming to hard lath martensite under deformation, particularly if the carbon content is low, conferring very high work-hardening characteristics and, potentially good wear resistance.

Metastable austenite/epsilon martensite Fe–Mn–Mo alloys have been developed [2], which were originally intended as a low-cost maraging-type of steel. They may find application in circumstances where wear resistance is important, such as for the mining industry, the agricultural sector, railway points and crossings.

These metastable austenite/epsilon Fe–Mn–Mo alloys produced good combinations of strength and toughness. These combinations are, of course, controlled not only by the chemical compositions of the alloys but also by their microstructures. The microstructure of metastable austenite/epsilon Fe–Mn–Mo alloys comprises austenite and epsilon with lath martensite in different proportions according to the chemical compositions and heat treatment. When a metastable Fe–Mn–Mo steel is deformed at room temperature it transforms martensitically from austenite/epsilon to hard lath martensite [3, 4].

The same effect has been noted in 304 stainless steels, which have an unstable microstructure and transform partially to martensite on deformation. These steels are hardened considerably by cold deformation. Other austenitic steels which are well known for giving a very high resistance both to impact and abrasion and possess a very high work-hardening capacity are the Hadfield manganese-type steels.

Explanations for the outstanding wear resistance of Hadfield steel were originally based on the strain-induced formation of martensite [5, 6], but it has subsequently been shown that austenite does not transform to martensite in this steel [7, 8] even below −196° C [9]. Furthermore, it is now known that the surface layers of the steel in which martensite was observed were, in fact, decarburized prior to deformation [9], and also local segregation of manganese, that leads to compositions which are unstable, can result in the observation of deformation-induced martensite. Sant and Smith [10] reconfirmed that deformation twinning is the primary mechanism for the work-hardening behaviour in Hadfield steel at room temperature.

The purpose of the present study was to determine the influence of composition and heat treatment on the strain-induced phase transformation and relate the phenomenon to the mechanical properties of these new Fe–Mn–Mo wear resistant/high-strength steels.

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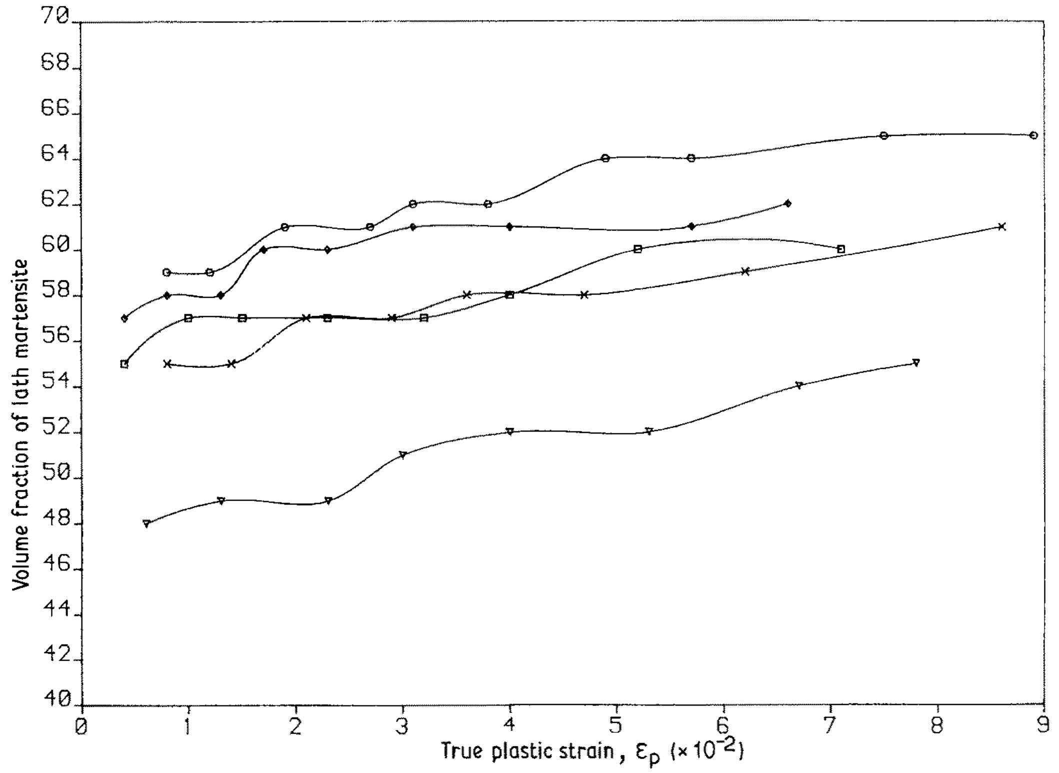

Figure 1 Volume fraction of lath martensite as a function of true plastic strain of the alloys which transformed during the mechanical testing in as-hot-rolled condition. (○) 0.05%, 11.30% Mn, 3.61% Mo; (◇) 0.020% C, 11.85% Mn, 2.65% Mo; (□) 0.07% C, 11.90% Mn, 1.91% Mo; (×) 0.025% C, 12.85% Mn, 2.70% Mo; (▽) 0.035% C, 13.47% Mn, 2.70% Mo.

2. Experimental procedure

2.1. Alloy preparation

A series of Fe–Mn–Mo alloys was prepared by vacuum induction melting so as to cover a wide range of structures by varying manganese (9 to 14%), molybdenum (2 to 4%) and carbon (0.001 to 0.1%) contents. Following melting and casting, the ingots were subjected to a homogenization treatment at 1200° C for 24 h. Subsequently, the ingots were heated to 1100° C and hot-rolled into 13 mm thick bar in five passes with about a 4% reduction in each pass. The finishing rolling temperatures were monitored by using an infrared pyrometer, and these are shown with the chemical analysis of the alloys in Table I. Each bar was machined into Hounsfield specimens for tensile testing and thereafter sealed in silica tubes under vacuum (to minimize carbon and manganese loss) for solution treatment at a range of temperatures (850, 950, 1050 and 1150° C) for 1 h and were then air-cooled.

2.2. Microstructural investigation

Representative specimens of heat-treated material were examined by optical microscopy. All the specimens were etched in nital or ferric chloride solution. The M_s temperatures of the alloys were determined by a dilatometer method on 5 mm thick specimens during continuous cooling from 1000° C. Details of the techniques are given elsewhere [11]. The approximate initial austenite/epsilon lath martensite contents were determined using a magnetic reluctance technique.

TABLE I Chemical analysis and the finishing rolling temperature

Alloys			Finishing temperature (° C)
C (%)	Mn (%)	Mo (%)	
0.003	8.60	1.99	900
0.05	11.30	3.61	862
0.020	11.85	2.65	862
0.070	11.90	1.93	817
0.025	12.85	2.70	850
0.035	13.50	2.70	850
0.060	13.60	3.43	892
0.003	14.10	2.07	851

2.3. Testing procedures

All tensile tests were carried out at a constant cross-head speed of (0.5 mm min^{-1}) corresponding to an initial strain rate of 8.33×10^{-3} sec^{-1}. Hardness measurements were made using a standard Vickers diamond pyramid testing machine with a 30 kg load. The extent of *in situ* transformation during the mechanical testing was monitored continuously by a magnetic reluctance technique. Full details of the techniques adopted are described elsewhere [11].

3. Results

The volume fraction of lath martensite before deformation was found to vary from 15% to 80% for the alloys under investigation. During deformation the amount of lath martensite was found to increase as the strain increased in the alloys that transformed, as shown in Figs 1 and 2. These curves can be compared with those of other workers such as Angel [12], Olson and Cohen [13] and Hecker *et al.* [14], who have studied strain-induced phase transformations.

3.1. Mechanical response

Examples of the true stress/true plastic strain curves and tensile work-hardening rate/true plastic strain curves obtained from the tensile tests are given in Figs

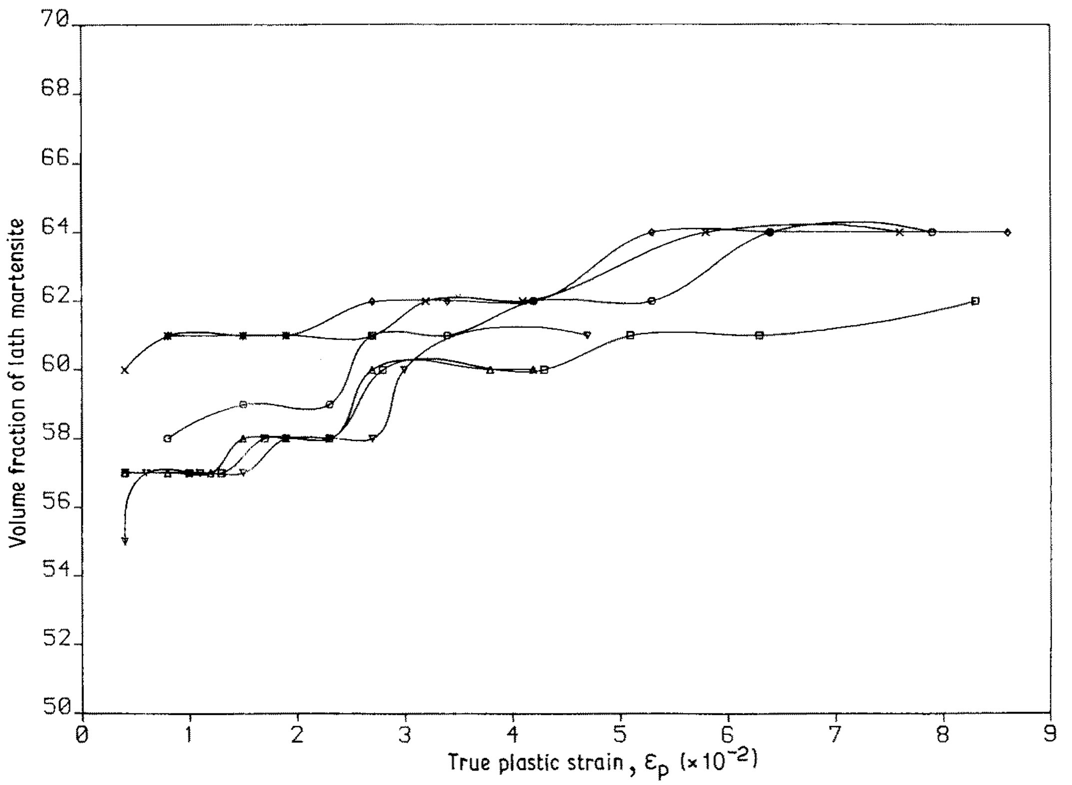

Figure 2 Volume fraction of lath martensite as a function of true plastic strain of the alloys which transformed during the mechanical testing in solution-treated condition. (○), (◇) 0.05% C, 11.30% Mn, 3.61% Mo; (□), (×) 0.020% C, 11.85% Mn, 2.65% Mo; (▽), (△) 0.025% C, 12.85% Mn, 2.70% Mo; (○, □, ▽) 850° C, (◇, ×, △) 950° C.

3 to 6. The overall flow stress levels and trends of the curves show how, as the transformation progressed, the work-hardening of the steels increased rapidly. The mechanical behaviour of a certain number of these steels is drastically changed when these alloys exhibit a phase transformation induced by plastic deformation (see Table II).

Recently, Inegbenebor *et al.* [4] have derived a composite flow stress equation for such an alloy undergoing deformation-induced phase transformation

$$\begin{aligned}\sigma_c = {} & \{(1 - V_{\alpha'})K[\ln(1 + \varepsilon)]^n V_\gamma^0 \\ & \times [1 - (1 + \varepsilon^{-s}/A)^{-1}][\varepsilon - \alpha V_{\alpha'}] \\ & + T[V_{\alpha'}^0 + V_\gamma^{0P}(1 + \varepsilon^{-s}/A)^{-P}[\varepsilon - \alpha V_{\alpha'}]\} \\ & \times [1 - \beta \, \mathrm{d}V_{\alpha'}/\mathrm{d}\varepsilon]\end{aligned}$$

where $V_{\alpha'}$ is the instantaneous volume fraction of lath martensite by the process of strain-induced transformation; K is the austenite strength factor, that is a measure of the capacity of the austenite to be strengthened by strain; ε the plastic strain; n, in this case, the austenite, strain-hardening index or work-hardening exponent; V_γ^0 the initial volume fraction of austenite; s the automotive lath martensite index; A the proportionality constant; α a constant related to dislocation behaviour; T the lath martensite strengthening factor; $V_{\alpha'}^0$ the initial volume fraction of lath martensite; P the lath martensite strengthening index; β a constant. This equation is based on a modified law of mixtures. From this composite flow stress equation the parameter s, i.e. the ability of lath martensite to accelerate the formation of additional lath martensite (e.g. [11, 15])

TABLE II Mechanical properties

Alloys			Elastic limit (MPa)	0.2% proof stress (MPa)	UTS (MPa)	Elongation (%)	n*	s
C (%)	Mn (%)	Mo (%)						
0.003	8.60	1.99†	500	549	882	16	0.319	–
0.05	11.30	3.61‡	333	364	1470	18	0.550	4
0.05	11.30	3.61†	312	338	1436	18	0.512	4
0.020	11.85	2.65‡	415	455	1276	20	0.643	4
0.020	11.85	2.65†	393	432	1304	19	0.580	4
0.070	11.90	1.93‡	422	472	1480	24	0.683	4
0.025	12.85	2.70‡	232	252	1333	23	0.457	4
0.025	12.85	2.70†	212	230	1172	10	0.410	4
0.035	13.50	2.70‡	160	197	1451	23	0.429	4
0.060	13.60	3.43†	119	138	1275	27	0.418	4
0.003	14.10	2.07†	89	101	814	33	0.218	4

*The austenite strain-hardening index or work-hardening exponent.

† Solution treated 950° C for 1 h.

‡ Hot-rolled condition.

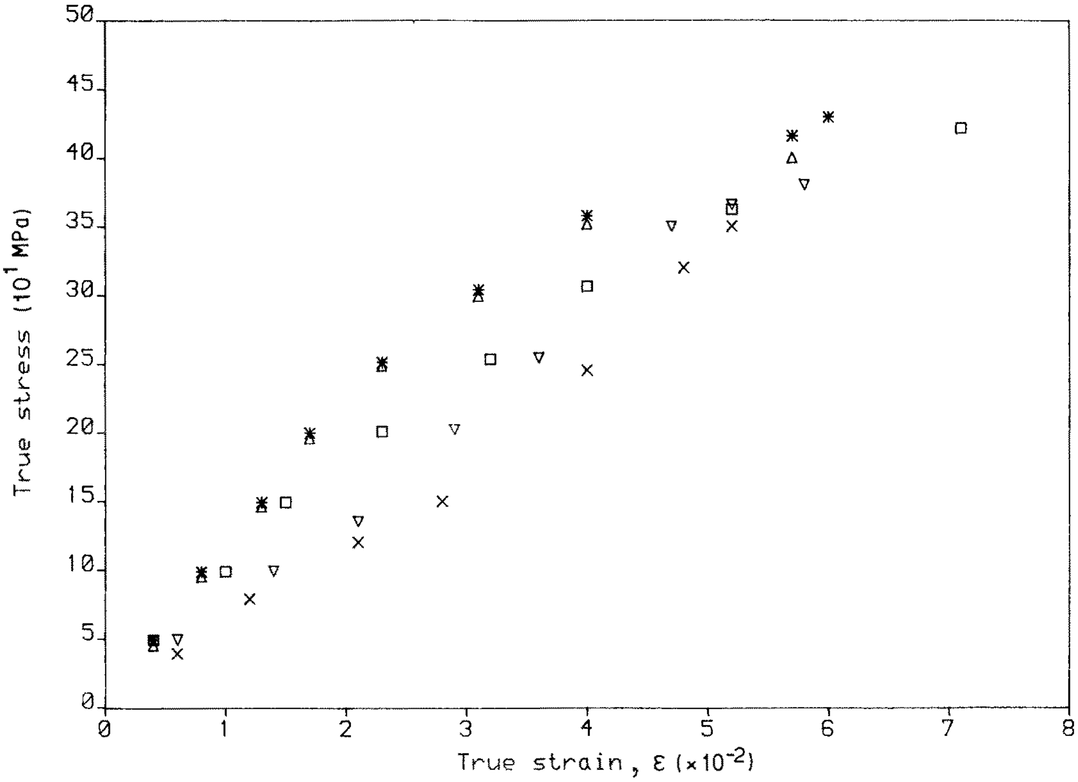

Figure 3 True stress plotted against true strain in as-hot-rolled condition. (△) 0.05% C, 11.30% Mn, 3.61% Mo; (∗) 0.020% C, 11.85% Mn, 2.65% Mo; (▽) 0.025% C, 12.85% Mn, 2.70% Mo; (×) 0.035% C, 13.50% Mn, 2.70% Mo.

has been derived. The values of this automotive lath martensite index for the alloys which transformed during the mechanical testing were closely grouped about "4". This finding suggests that the automotive lath martensite aspect of strain-induced transformation is insensitive to composition and conditions of treatment.

3.2. Metallographic observations

The structure of the as-solution-treated alloys which did not transform, and the ones which transformed during the mechanical testing were examined by optical microscopy and representative micrographs of the specimens are given in Figs 7 and 8. In Fig. 8, long straight bands of lath martensite are observed.

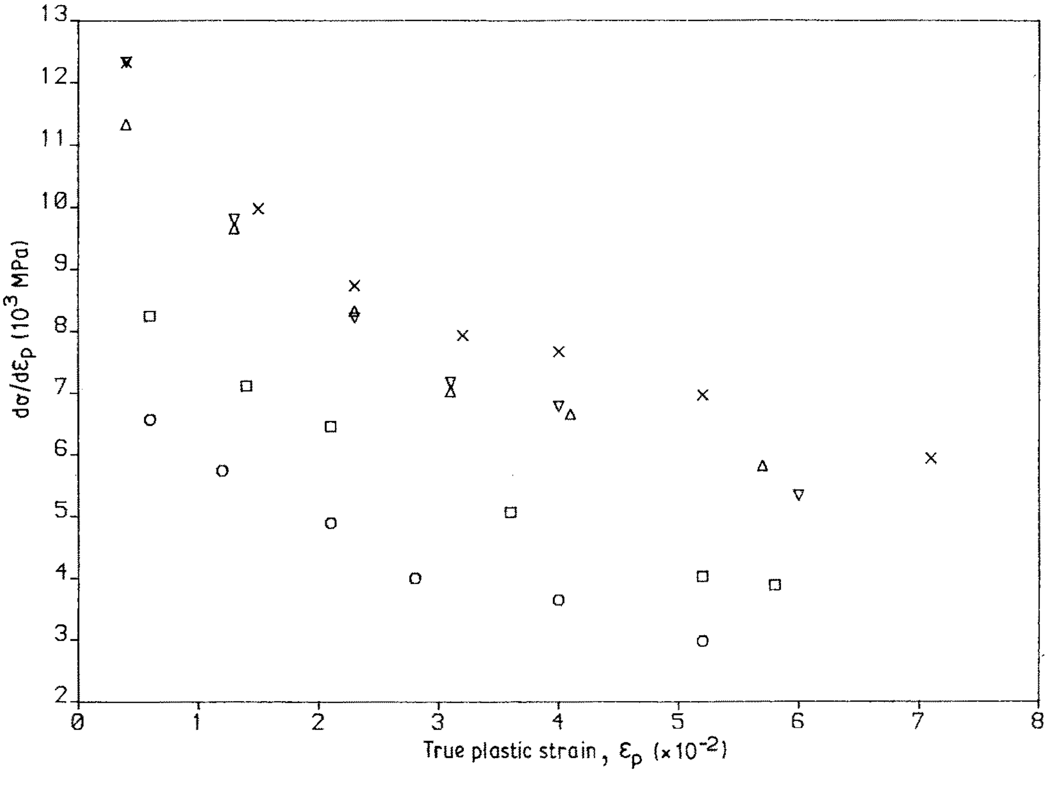

Figure 4 Tensile work-hardening rate plotted against true plastic stress in as-hot-rolled condition. (△) 0.05% C, 11.30% Mn, 3.61% Mo; (▽) 0.020% C, 11.85% Mn, 2.65% Mo; (×) 0.07% C, 11.90% Mn, 1.93% Mo; (□) 0.025% C, 12.85% Mn, 2.70% Mo; (○) 0.035% C, 13.5% Mn, 2.70% Mo.

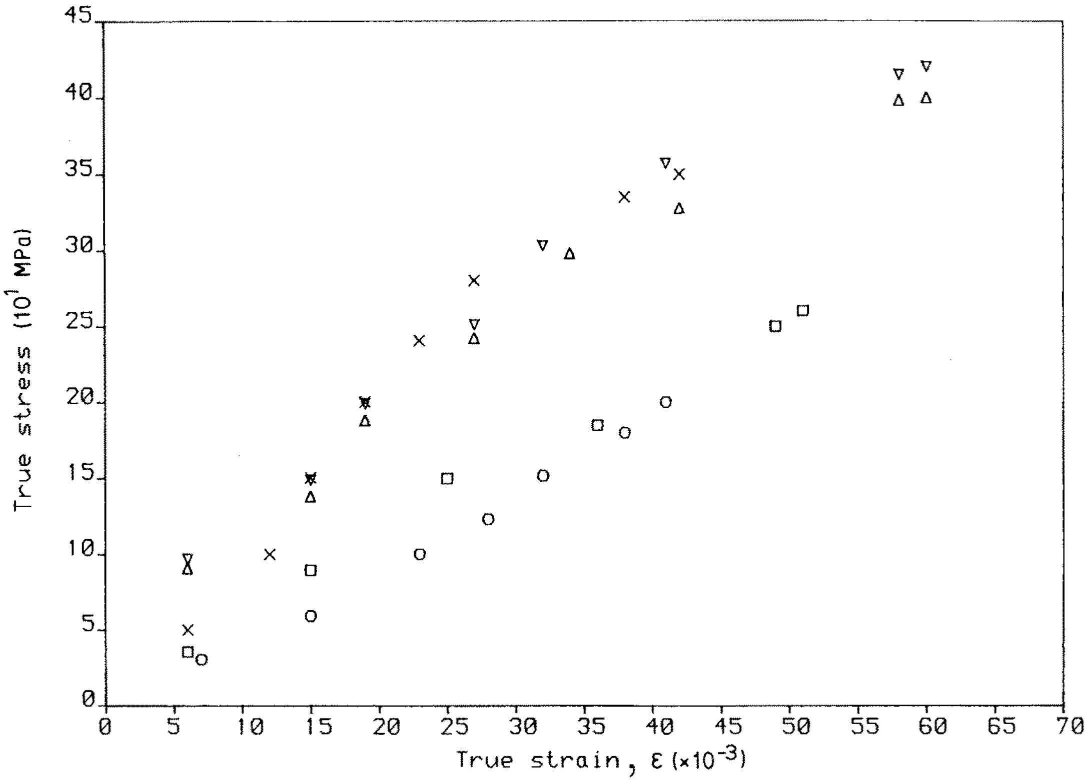

Figure 5 True stress plotted against true strain in solution-treated condition. (△) 0.05% C, 11.30% Mn, 3.61% Mo; (▽) 0.020% C, 11.85% Mn, 2.65% Mo; (×) 0.025% C, 12.85% Mn, 2.70% Mo; (□) 0.06% C, 13.6% Mn, 3.43% Mo; (○) 0.003% C, 14.10% Mn, 2.07% Mo.

4. Discussion

The compositions investigated gave microstructures ranging from a fully martensitic state to unstable austenite/epsilon-lath martensite. The M_s temperatures of all the alloys were measured. The M_s temperatures of the unstable austenite/epsilon alloys were above room temperature whilst their M_f temperatures were below room temperature. During mechanical testing, these unstable austenite/epsilon alloys transformed to hard lath martensite on straining indicating that their M_d temperatures were above their M_s temperatures (M_d is defined as the temperature – in practice usually a range – at which the martensite transformation takes place during plastic deformation). The transformation of the unstable austenite/epsilon to lath martensite during straining provided exceptionally high work-hardening rates as shown in Figs 3 and 4. The irregular nature of the curves in Figs 1 and 2

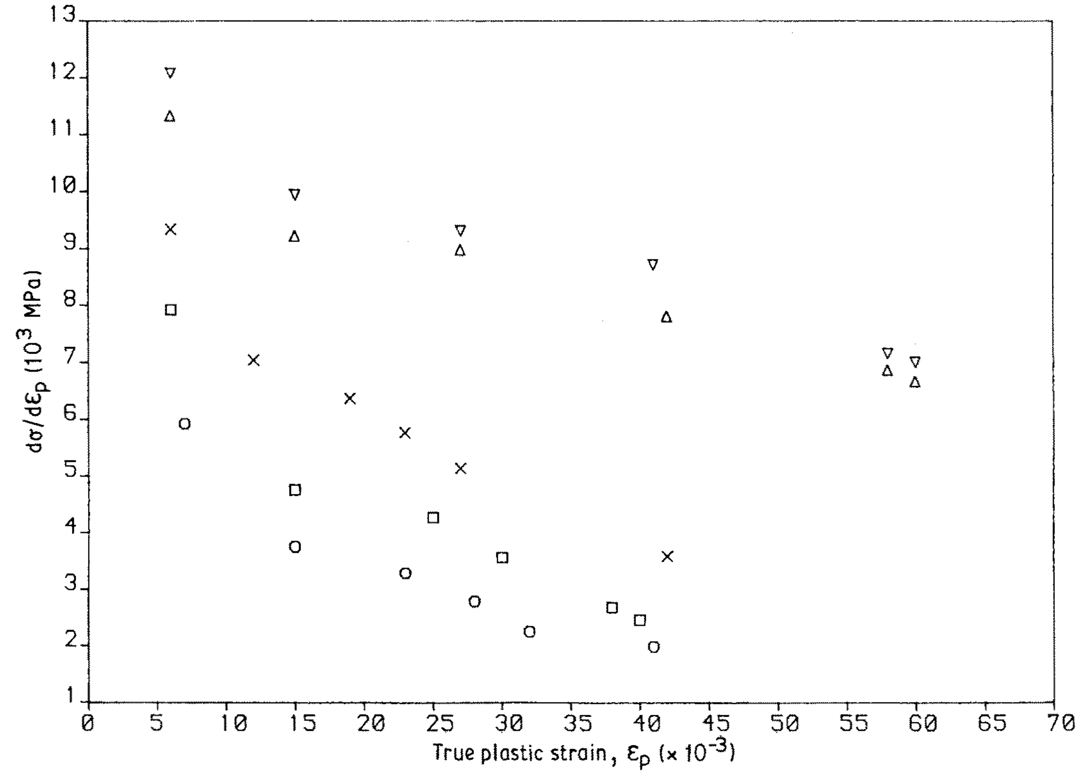

Figure 6 Tensile work-hardening rate plotted against true plastic stress in solution-treated condition. For key, see Fig. 5.

Figure 7 Optical micrograph of the alloy which does not transform during mechanical testing, solution treated at 950° C for 1 h and air cooled.

Figure 8 Optical micrograph of the alloy which transformed during mechanical testing, solution treated at 950° C for 1 h and air cooled.

seems to suggest a transformation "burst" phenomenon [16] during plastic deformation. This might be associated with the "brickwork" martensitic structure which forms in these steels as a result of the fragmentation of the original austenite on (1 1 1) planes by thin layers of epsilon phase. This (i.e. epsilon) is an intermediate in the strain-induced transformation to lath martensite as reported by Bogachev *et al.* [17] and Jones *et al.* [2]. What happens is that the nucleation of one lath triggers off the formation of another lath in an adjacent region as plastic strain increases. This statement might be taken as a justification for the values of the automotive lath martensite index *s* of all the alloys which transformed being similar (~ 4), that is, the formation of automotive lath martensite is independent of composition (over a range) and the conditions of heat treatment. However, the automotive nature of the formation of lath martensite depends on the strains or stresses surrounding existing lath martensite regions which trigger off its formation during the strain-induced transformation.

In the model of Olson and Cohen [18], formation of martensite by plastic deformation and how this is affected by stacking fault energy and strain rate has been explained. During a strain-induced transformation, plastic deformation of the parent phase creates the correct defect structures to act as embryos for the transformation products. This will result in a higher dislocation density in the end product and an increased stacking fault population in the austenite/epsilon matrix. Thus, the exceptional work-hardening rates that have been achieved in the low-strain region in these unstable austenitic steels is due to strain-induced formation of fine, heavily dislocated lath martensite and also probably to an increase in stacking faults in the surrounding austenite as a result of the strain-induced transformation.

To sum up, an unstable austenite/epsilon martensitic microstructure is obtained in alloys containing 11 to 14% Mn, 2 to 4% Mo and 0.1% C. These steels have good proof stresses (surpassing Hadfield steel), high tensile strength, reasonable ductility, exceptional work-hardening characteristics and good toughness. These properties are appropriate for wear-resistant applications.

5. Conclusions

1. The microstructures found depend mainly on the manganese contents: the fully martensitic alloys contain 9% Mn while unstable austenite/epsilon martensite is found in the range from about 11 to 14% Mn.
2. The mechanical properties of these unstable austenite/epsilon martensite steels were very good in both the as-hot-rolled condition and after solution-treatment at 950° C.
3. All the unstable austenite/epsilon steels transformed to lath martensite during straining, giving exceptional work-hardening characteristics and this work-hardening capacity exceeded that of Hadfield manganese steel.
4. The exceptional work-hardening characteristics of these alloys are attributed to dislocation strengthening, an increase in stacking fault population in the matrix and formation of strain-induced lath martensite.

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