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To cite this version:
M. Andersson, J. Ågren. Effect of Prestraining and Training on the $\gamma \rightarrow \epsilon$ Transformation in Fe-Mn-Si Alloys. Journal de Physique IV Colloque, 1995, 05 (C8), pp.C8-457-C8-462. <10.1051/jp4:1995869>. <jpa-00254118>

HAL Id: jpa-00254118
https://hal.archives-ouvertes.fr/jpa-00254118
Submitted on 1 Jan 1995

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Effect of Prestraining and Training on the $\gamma \rightarrow \epsilon$ Transformation in Fe-Mn-Si Alloys

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Abstract. The effect of training and predeformation on the martensitic transformation is investigated. It is concluded that $M_s$ as a function of predeformation strain must have a maximum at around 3% strain. Prestrains less than 5% will enhance the martensitic transformation whereas larger prestrains depress it due to strain hardening of the $\gamma$ phase.

1. Introduction

In some Fe-Mn-Si alloys a martensitic transformation, $\gamma \rightarrow \epsilon$, can be induced by straining. If $\epsilon$ martensite is induced during straining a shape memory effect (SME) is observed upon heating to temperatures where the reverse transformation $\epsilon \rightarrow \gamma$ occurs. The shape memory behaviour is due to the reversible movement of $[a/6][112]$ Schockley partial dislocations during straining and shape recovery annealing [1, 2, 3].

After the work by Sato et al. [1], who first reported SME in Fe-Mn-Si single crystals, much research have been performed in order to improve the SME in this type of alloys. In 1986 Murakami et al. [4] reported a complete shape memory effect in polycrystalline Fe-Mn-Si alloys with 28 to 34 w% Mn and 4 to 6 w% Si. The same research group also found that the SME could be improved by a repetition of a small deformation at room temperature followed by an annealing at 773-973 K. Such a treatment has been called training [5]. The training makes the austenite matrix stronger against slip and lowers the critical stress for martensite formation [6]. The combination of these two effects decreases the amount of platic deformation caused by dislocation glide and increases the amount caused by martensitic transformation and consequently the SME may be improved by training. Many investigations, [6, 7], have been performed to evaluate the optimal training conditions, i.e. to find the right deformation temperature, strain and annealing temperature for different materials. Several researchers have also investigated why the training improves the SME.[6, 8, 9]

The purpose of the present report is to analyse further the effect of training on the martensitic transformation $\gamma \rightarrow \epsilon$ in Fe-Mn-Si alloys. In particular we shall analyse how the martensitic transformation is influenced by predeformation of the austenite above $M_S^d$, i.e. the highest temperature where martensite may be induced by stress or strain.

2. Experimental procedure

The alloy was prepared by induction melting under argon atmosphere and cast in a copper mould. The alloy composition is given in table 1.
Table 1. The chemical composition (in weight percent) of the alloy.

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<tbody>
<tr>
<td>Fe</td>
<td>Mn</td>
<td>Si</td>
<td>C</td>
<td>S</td>
<td>N</td>
</tr>
<tr>
<td>Bal.</td>
<td>30.67(±0.05)</td>
<td>4.95(±0.05)</td>
<td>0.007</td>
<td>0.009</td>
<td>0.008</td>
</tr>
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</table>

The alloy was annealed 1.5 h at 1373 K, in argon gas atmosphere, before it was reduced 50% by hot rolling. Thereafter it was homogenized at 1273 K for 24 h.

Tensile test specimens with a diameter of 3 mm and a gavelength \( l_0 = 20 \) mm were machined from the homogenized alloy. The test specimens were finally annealed for 10 min at 1173 K to remove any deformation that could have been induced during machining. The tensile tests were performed on an Instron machine with a crosshead speed of 0.5 mm/min.

Tensile tests of 3% strain were performed at room temperature and the samples were subsequently heat treated in two different ways:

A: Annealing in a salt bath for 15 min at 873 K and followed by air-cooling to room temperature.

B: Heating in a dilatometer, 5 K/min, followed by an isothermal treatment for 15 min at 873 K and thereafter cooling inside the furnace to room temperature.

The SME in percent is given by:

\[
\text{SME} = 100 \times \frac{(l_1 - l_2)}{(l_1 - l_0)}
\]  

where \( l_0 \) is the gavel length before deformation, \( l_1 \) after deformation and \( l_2 \) after heating to 873 K.

The transformation temperatures, \( M_S = 293 \) K, \( A_S = 417 \) K and \( A_F = 432 \) K, were measured in the undeformed material by means of DSC and the slope baseline intersection method.

3. Results

The results from the training treatment is plotted in figs. 1.a-b. Fig. 1.a shows the variation in SME and fig. 1.b the variation in \( \sigma_{0.2} \) and \( \sigma_{2.5} \) stresses, i.e. the stresses at 0.2 and 2.5 % plastic deformation respectively, as a function of number of training cycles. The dashed and solid lines show the results from samples heat treated according to method A and B respectively. As can be seen, the two heat treatment methods do not yield any important difference in the final result.

The SME increases rapidly in the first few training cycles and reaches a constant value of 90 % already after 3 cycles, fig. 1.a. The \( \sigma_{0.2} \) decreases from 217 MPa, in the first cycle, to 126 MPa, in the second cycle, fig. 1.b. Further cycling gives no further decrease and \( \sigma_{0.2} \) seems to have approached a constant value of 130 MPa. The \( \sigma_{2.5} \) shows a similar variation, although less strong, and decreases form 360 MPa in the first cycle to 320 MPa in the following cycles.
4. Discussion

The experimental results show a large increase in the SME after the first training cycle, see fig. 1a, and a large decrease in yield stress, see fig. 1b. This behaviour is in agreement with previous investigations in Fe-Mn-Si based alloys [10, 11, 12].

Upon further cycling the increase in SME is much smaller and the yield stress is hardly affected at all. The decrease in yield stress can be interpreted as a decrease in the stress needed to induce \( \varepsilon \) martensite during the deformation.

During each deformation step the applied strain can be divided into two parts, the plastic strain of the austenite due to dislocation glide, \( \varepsilon_p \), and the transformation strain, \( \varepsilon_t \), due to the martensitic transformation. During the heating step of a training cycle there is no recrystallization although there will be some recovery, i.e. a fraction, but not all, of the dislocations generated during the slip will be annihilated. After each training cycle the dislocation density of the austenite matrix thus increases and there is deformation hardening of the austenite. The training treatment should thus be equivalent with a predeformation of the austenite above \( M_S^d \).

The effect of austenite predeformation above \( M_S^d \), have been studied by Tsuzaki et al. [13], who considered the \( \gamma \rightarrow \varepsilon \) martensitic transformation in Fe-Mn alloys. They showed that the \( M_S \) temperature monotonously decreases with increasing deformation of the austenite in the range 6-40\%. They also suggested that this effect could be quantitatively accounted for by the model of Olson and Cohen [14] assuming that an extra driving force is needed in order to overcome the friction for dislocation glide. The extra driving force was assumed proportional to the increase in yield stress caused by the strain hardening.

In a more recent article Tsuzaki et al. [8] showed that the \( \sigma_{0.2} \) in Fe-Mn-Si alloys is lower in alloys strained 0-5% at temperatures above \( M_S^d \) than in an undeformed material. The most reasonable
interpretation is that the small prestraining allows a subsequent deformation to proceed with a martensitic mechanism rather than dislocation glide. The decrease in $\sigma_{0.2}$ should thus be due to a decrease in the driving force needed to form martensite. Such a decrease should yield an increase in $M_s$. According to the Olson and Cohen model martensite would nucleate when

$$\frac{nd}{V_m}(\Delta G_m + \Delta G_{el}^m + \Delta G_{\tau}^m) + 2\gamma = 0 \quad (2)$$

where $n$ is the number of close packed planes that comprise the nucleus, i.e. the defect size and $d$ the spacing between the close-packed plane, $V_m$ the molar volume, $\gamma$ the interfacial energy. $\Delta G_m$, $\Delta G_{el}^m$ and $\Delta G_{\tau}^m$ are, respectively, the thermodynamic driving force (negative for a reaction to occur), the elastic energy and the driving force needed to overcome the glide resistance caused by strain hardening. The subscript $m$ signifies that all the $\Delta G$:s are counted per mole of atoms. Tsuzaki et. al. assumed

$$\Delta G_{\tau}^m = V_m \left( \frac{b}{md} \right) \tau \quad (3)$$

where $m=2$ for the $\gamma \rightarrow \epsilon$ transformation, $b$ the Burgers vector and $\tau$ the strainhardening. As can be seen from eq.2 a slight increase in $\Delta G_{\tau}^m$ could be compensated for by an increase in $n$. The experimental information on the yield stress [8] thus suggests that low prestrains increase the potency of the defects and make the martensitic transformation simpler. As the prestrain increases $\Delta G_{\tau}^m$ becomes dominating and nucleation can only occur if the driving force increases, i.e. the yield stress increases and $M_s$ decreases.

From the predeformation experiments we may thus conclude that $M_s$ as a function of predeformation strain should have a maximum at approximately 3%, see schematic diagram in fig. 3. Below 5% strain the martensitic transformation is enhanced and $M_s$ is higher than in the undeformed material. Above 5% strain the depressive effect from strain hardening dominates and $M_s$ is lower than in undeformed material. This does not necessarily mean that the SME would be worse at high prestrains because the yield stress for dislocation glide is still substantially higher than that for martensite formation. It simply means that the yield stress of the alloy would be higher.
This result suggests that one should not use much larger strains than 5% in the training process unless one wants a higher yield strength. It also seems as a low strain increment in the training cycle would eventually, after many cycles, lead to a depression of the martensitic transformation provided that SME is below 100%.

As discussed by Tsuzaki et al.[8], the SME does not only depend on the condition that the deformation should proceed by martensitic transformation rather than dislocation glide but also on the reversibility of the martensitic interface itself. Tsuzaki et al. suggested that the back stress formed by the martensitic transformation may play a role and if it is relaxed by dislocation glide the reversibility would be low, yielding a poor SME.

5. Conclusions

The main purposes of the training is to strengthen the austenite in order to reduce the amount of plastic deformation by dislocation glide and to make martensite formation easier. However large prestrains would actually depress the martensite formation and it is suggested that $M_s$ as a function of predeformation strain should have a maximum at approximately 3%. The straining increment during a training cycle should thus be around 3% unless one wants a higher yield stress of the alloy.

Acknowledgements

This project is financed by the Swedish National Board for industrial and Technical Development.

References

