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Microstructure–property correlation in martensite–austenite mixtures*

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The mechanical response of several martensite–austenite mixtures was analysed and related to the existent microstructures. It was found that the mixtures behave as coarse dispersions with deformation taking place by and large in the matrix, austenite. The connectivity of austenite was found to be the crucial factor determining uniform strain.

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1. Introduction

The rationalization of the mechanical behaviour of multiphase materials is of considerable interest since the majority of engineering materials comprise such complex mixtures. According with Hornbogen's views [1] two-phase mixtures can display three basic types of microstructure which allows their classification into three classes: dispersion, net and duplex.

A dispersion contains the second phase well mixed in a continuous matrix. A net displays the second phase as a skeleton along the grain boundaries of the matrix which, consequently, is not continuous. A duplex structure may be described as a coarse dispersion in which both precipiate and matrix are only partially continuous.

In this work, consideration is given to the tensile properties of martensite–austenite mixtures in Fe–31.9% Ni–0.02% C. A typical microstructure is shown in Fig. 1. Clearly it is more complex than the prototypes identified and analysed by Hornbogen and co-workers [1, 2]. Nonetheless, the material was selected for the present investigation based on the following considerations. The alloy is stable at room temperature and the microstructure of the mixtures can be easily described by optical microscopy [3, 4]. The mechanical data for both austenite and martensite can be obtained independently of each other. The production of different volume fractions does not involve changes in chemical composition of the martensite, such as in dual phase steel [5]. Therefore it appears that our alloy constitutes an appealing system for investigating basic aspects of the mechanical response of two-phase mixtures.

2. Experimental methods

The material, a high purity Fe–31.9% Ni–0.02% C alloy (nominal composition) was received in the form of a flat ingot about 25.4 mm thick. This material was cut into suitable pieces and machined to remove a layer of about 1 mm. These pieces were annealed at 1373 K for 12 h under vacuum inside quartz capsules, cold-rolled to 1 mm strips and quenched into liquid nitrogen. These strips were machined into flat tensile specimens with a reduced gauge length (27 mm × 4 mm × 1 mm) and given a final vacuum annealing at 1473 K inside quartz capsules. These austenitic specimens with a 64.4 μm grain size were cooled by immersion into liquid nitrogen refrigerated baths to obtain different volume fractions of martensite. Some pieces were annealed at lower temperatures (1173, 1273 and 1373 K) to produce austenite specimens with smaller grain sizes.

Quantitative microscopy was used to describe the microstructure of each specimen. Three perpendicular cross-sections were observed under a

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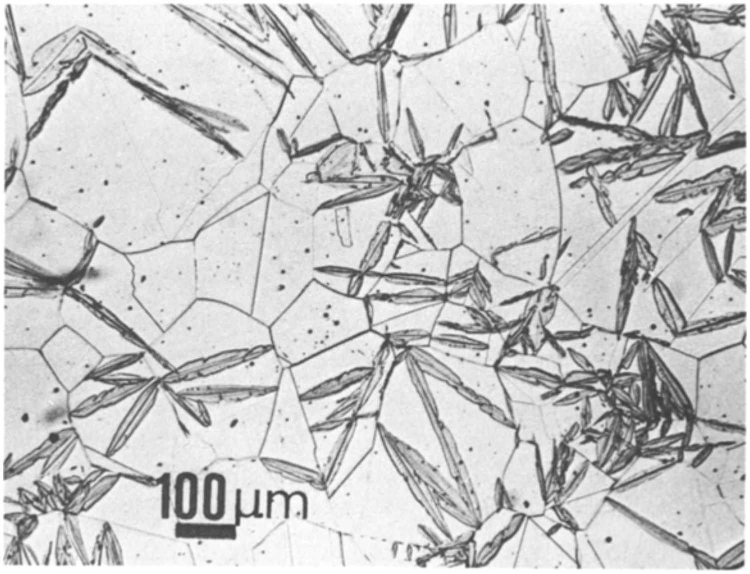

Figure 1 Optical photomicrograph showing typical microstructure of partially transformed Fe–31.9% Ni–0.02% C.

Leitz Orthoplan microscope equipped with suitable eye-pieces. Volume fractions were obtained by point-counting. Lineal analysis was used to obtain grain size and related parameters such as the area of interphase boundaries per unit volume of material. Systematic sampling was applied and the data (mean values) herein reported have a statistical error of about ± 10% with 95% confidence.

The mechanical testing was performed with a TT-DM Instron machine at room temperature and the data read from the chart of the machine were converted into stress and strain values with the aid of a computer. A nominal strain rate of 10^{-4} sec^{-1} was used in all tests.

3. Experimental results

3.1. Microstructure

The photomicrograph of Fig. 1 is typical of a specimen with a martensite volume fraction $V_\text{v} = 0.31$. Examination of several specimens disclosed that the transformation is well spread throughout the grains as expected in view of the coarse austenite grain size obtained with the heat treatment applied [3]. The evolution of the microstructure of the material with the progress of the transformation followed the same trend described in earlier publications [3, 4]. Table I contains the values of the basic parameters determined by point-counting and by lineal analysis.

The mechanical properties of two-phase materials are related to the amount and distribution of the phases. In order to estimate these, connectivity or continuity parameters [5] were computed for each phase,

$$C^\gamma = \frac{2S_\text{v}^{\gamma\gamma}}{2S_\text{v}^{\gamma\gamma} + S_\text{v}^{\alpha\gamma}} \tag{1}$$

$$C^\alpha = \frac{2S_\text{v}^{\alpha\alpha}}{2S_\text{v}^{\alpha\alpha} + S_\text{v}^{\alpha\gamma}} \tag{2}$$

where S_v stands for the area of interface per unit volume and the superscripts α and γ stand for martensite and austenite, respectively. Fig. 2 shows the variation of C^γ and C^α as functions of the volume fraction transformed. Two facts emerge from this. First, it is clear that both phases are only partially continuous. Moreover their connectivities are not too different for values of V_v from 0.21 to 0.41. Second, the connectivity of martensite appears not to vary significantly with the fraction transformed up to $V_\text{v} = 0.41$, whereas the connectivity of austenite

TABLE I Metallographic data. S_v represents area of interface per unit volume of material. The superscripts γ and α represent austenite and martensite, respectively. V_v is the martensite volume fraction

Martensite volume fraction V_v	Ratio of interphase area $S_\text{v}^{\gamma\text{-}\alpha}/S_\text{v}^{\alpha\text{-}\alpha}$
0.28	4.2
0.33	4.1
0.41	4.2
0.62	2.9
0.73	1.8

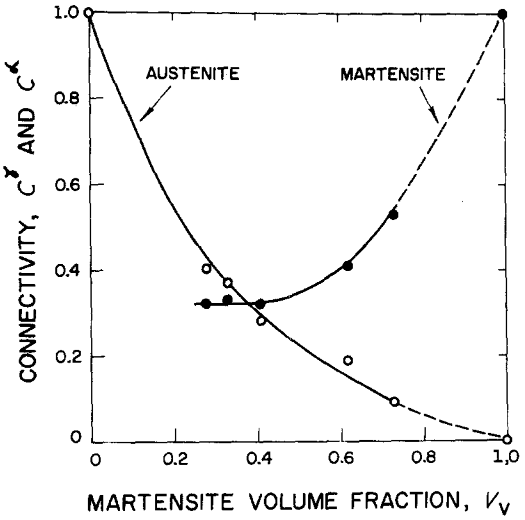

Figure 2 Variation of the connectivity of austenity, C^{γ}, and of martensite, C^{α}, as a function of the volume fraction transformed.

decreases continuously. The former may be attributed to autocatalysis and the latter to the fact that in these high-nickel, carbon alloys which transform to plate martensite, the plates partition the austenite grains impinging on the austenite boundaries. This phenomenon can be described by the variation of the mean free distance in the austenite, λ, with increasing V_V as shown in Fig. 3.

3.2. Tensile properties

The curves of Fig. 4 show the tensile response at room temperature of two mixtures (one with 33% and the other with 73% martensite) and those of pure austenite and martensite. The latter was obtained with a specimen previously cooled into liquid helium. Although it was not possible to determine exactly the amount of retained austenite in that material, it was estimated to be considerably less than 0.05%.

The data show that increasing the martensite content results in mixtures with higher ultimate tensile strength (UTS), smaller total elongation values and enhanced hardening at low strains. Moreover, the data indicate that the martensite is clearly stronger than the austenite. Their flow stresses at proof strain of 0.2% are in the ratio of 2.4:1. In addition the UTS of austenite is less than the 0.2% flow stress of martensite. Thus, to a first approximation, the mixtures can be considered to be constituted of a soft matrix and a hard reinforcing phase.

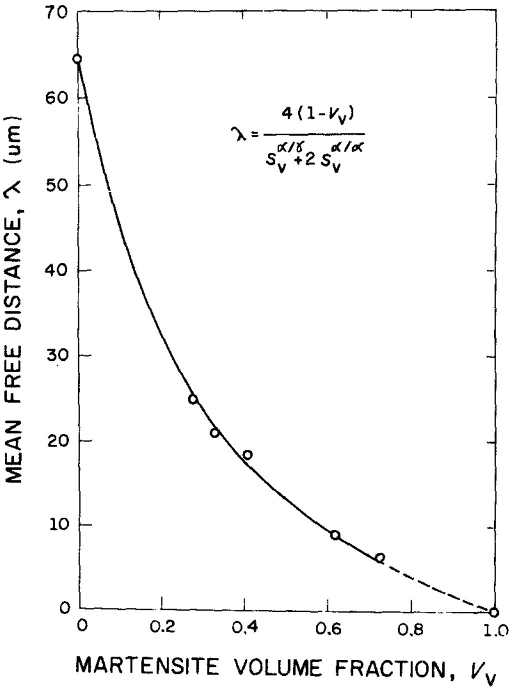

Figure 3 Variation of the mean free distance in the austenite, λ, as a function of the volume fraction transformed.

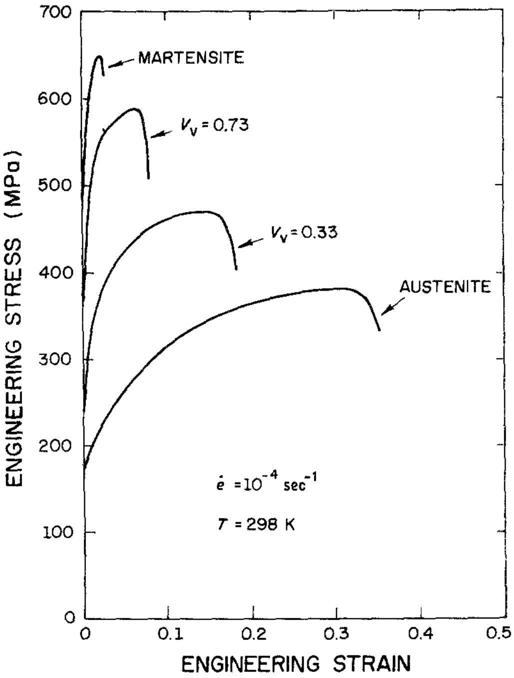

Figure 4 Stress–strain curves of two mixtures and those of pure austenite and pure martensite.

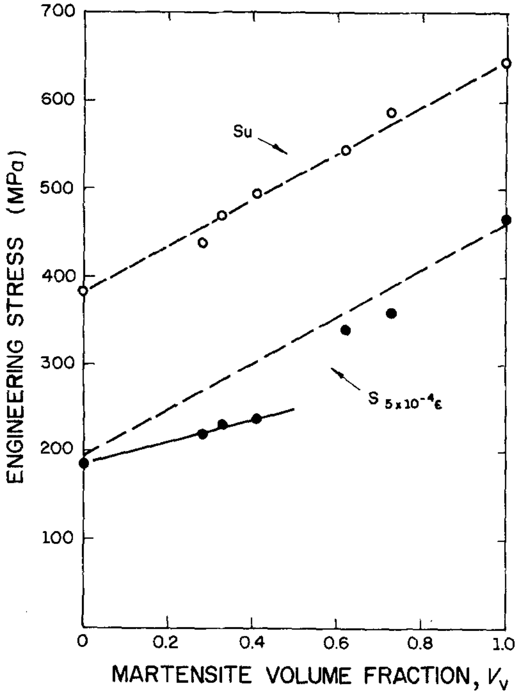

Figure 5 Rule of mixtures plots for 0.2% proof stress and UTS. Unfilled symbols represent austenite data.

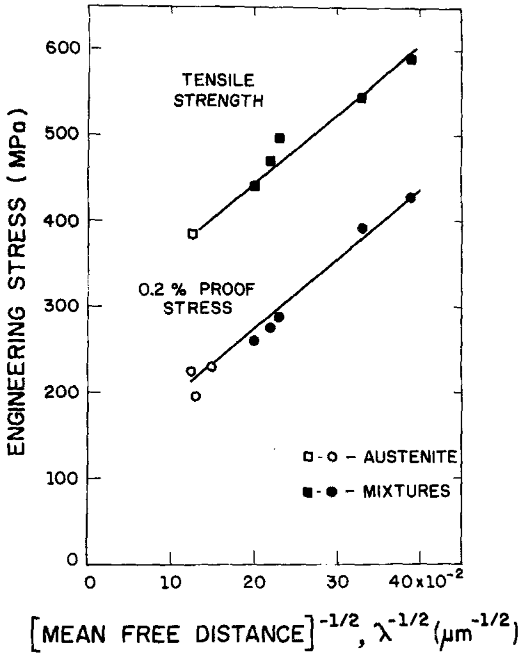

Figure 6 Hall–Petch plots for 0.2% proof stress and UTS. Unfilled symbols represent austenite data; filled symbols represent mixtures.

Regarding the influence of martensite volume fraction, the data of Fig. 5 indicate that the mechanical strength, UTS, of the mixtures is directly proportional to the martensite content. A simple "rule of mixtures" might be considered to explain this result. However it is only apparently so because the strain of the mixtures at UTS are different. This contention is confirmed by the fact that the proof stress values at 0.2% strain, also shown in Fig. 5, deviate from the line which represents the "rule of mixtures". These observations are in qualitative agreement with the results of Tamura *et al.* [6, 7] who proposed that the difference in the yield strengths of the two constituents would cause deviations from the "rule of mixtures".

3.3. Microstructure–property correlation

According to Gurland [8] a mixture comprising discontinuous constituents should yield when one of them becomes plastic. As the austenite is weaker than the martensite, it is proposed that in the mixtures of concern the matrix (austenite) controls yielding.

In order to verify the validity of the above proposition, the values of the flow stress at 0.2% proof strain, typical of mixtures and of untransformed austenitic specimens were plotted after the well known Hall–Petch relation, Fig. 6. In the case of mixtures, the mean austenitic free distance, λ, was substituted for the austenite grain size. The 0.2% proof strain was selected to dilute any influence from substructure introduced by the transformation itself. Inspection of Fig. 6 indicates that this set of data may be described by a single line. Thus, the Hall–Petch plot supports the previous assumption.

Insight on the work-hardening of the mixtures was also gained by considering the Hall–Petch plot of the UTS data (represented by squares in Fig. 6). Once more, a single line can be used to describe the data. However, it is striking that this line is essentially parallel to that through the yield data. Thus, it appears that the stress increment possible by work-hardening the mixtures does not depend upon the martensite volume fraction in the material. In fact it is the same as that obtained with pure austenite. This provides further support to the contention that the austenite is the major deforming phase in the mixtures.

Then, assuming that the strain in martensite is null, and invoking the principle of partial mechanical properties advanced by Hornbogen [2], it is possible to write the following expression for the

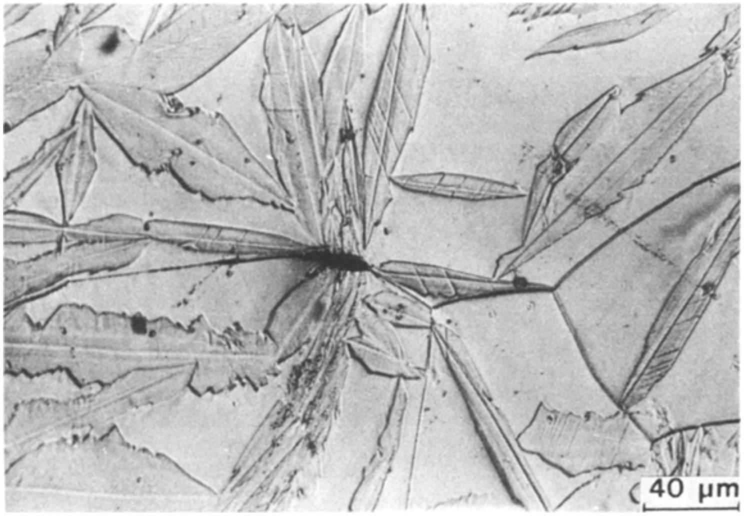

Figure 7 Optical photomicrograph showing the microstructure of a mixture with $V_v = 0.41$ deformed to fracture. Notice a grain boundary crack near the sites where martensite impinges. Observe that the austenite/martensite boundaries are intact.

uniform strain e_u^* of the mixtures

$$e_u^* = (1 - V_v)\,\bar{e}_\gamma$$

where $\bar{e}_\gamma$ is the effective strain in the austenite and V_v is the martensite volume fraction. The computed values of $\bar{e}_\gamma$ are shown in Table II. It is clear from the table that, with only one exception, the values obtained for $\bar{e}_\gamma$ are approximately constant, but smaller than the uniform strain of pure austenite. The constancy of $\bar{e}_\gamma$ is clearly consistent with the Hall–Petch analysis. However, an explanation is required for $\bar{e}_\gamma$ being less than the uniform strain of pure austenite.

Observation of several specimens deformed to fracture did not reveal signs of decohesion along the martensite–austenite interfaces. However cracks were visible at the austenite grain boundaries where martensite impinged, Fig. 7. Regarding this, it is noteworthy that the measured uniform strains could be related to the connectivity of the austenite, C^γ, by a "rule of mixtures" with C^γ substituted for volume fraction, Fig. 8. Hence, it appears reasonable to conclude that in the mixtures the austenite displays less uniform deformation than is possible when untransformed, due to a reduced capacity for deformation (compatibility) along the grain boundaries obliterated by martensite.

TABLE II Values of strain

Experimental data		Calculated data
V_v	e_u^* (uniform strain)	$\bar{e}_\gamma$ (effective strain)
0	0.309	0.309
0.28	0.157	0.253
0.33	0.161	0.240
0.41	0.133	0.225
0.62	0.063	0.166
0.73	0.067	0.248

In summary, the analysis above allows the conclusion that in austenite–martensite mixtures containing plate martensite, the latter is by and large providing dispersion hardening. However, as the martensite–austenite interfaces proved to be resistant to decohesion, it is also possible that fibre reinforcement (by load transfer) may be taking place.

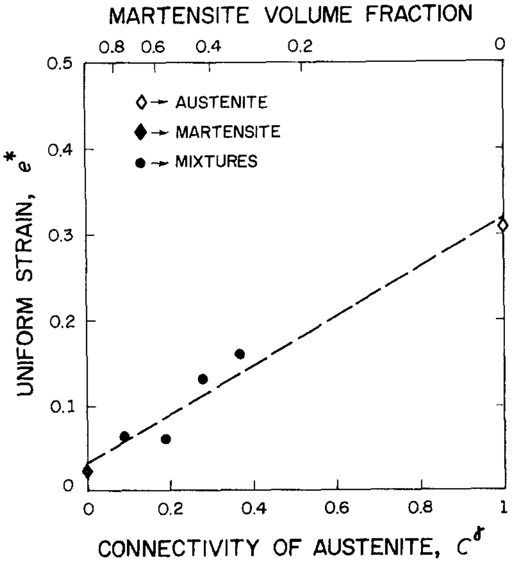

Figure 8 Variation of uniform strain as a function of the connectivity of austenite. Datum for martensite ($V_v = 1$) was included for comparison.

4. Conclusions

1. Mixtures comprising plate martensite in an austenite matrix behave as coarse dispersions.

2. Deformation takes place by and large in the austenite and the major strengthening effect of martensite is obtained through the reduction of the effective austenite grain size due to the partitioning of the austenite grains by the martensite plates.

3. The ductility of the mixtures in uniform elongation is controlled by the connectivity of the matrix.

4. The sites of impingement of martensite plates with the austenite grain boundaries are preferred sites for fracture initiation.

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References

1. J. BECKER and E. HORNBOGEN, in "Structure and Properties of Dual-Phase Steels", edited by R. A. Kot and J. W. Morris (TMS-AIME, Warrendale, 1979) p. 20.
2. E. HORNBOGEN, in "Strength of Metals and Alloys" Vol. III, edited by R. C. Gifkins (Pergamon Press, London, 1983) p. 1059.
3. J. R. C. GUIMARÃES and J. C. GOMES, *Acta Metall.* **26** (1978) 1591.
4. *Idem, Met. Trans.* **10A** (1979) 109.
5. G. R. SPEICH and R. L. MILLER, *ibid.* **10A** (1979) 154.
6. I. TAMURA, Y. TOMOTA, A. AKAO, Y. YAMAOKA, M. OZAWA and S. KANATAN, *Trans. ISIJ* **13** (1973) 283.
7. I. TAMURA, Y. TOMOTA and M. OZAWA, Proceedings of the 3rd International Conference on the Strength of Metals and Alloys, Cambridge (1973) Vol. I, p. 611.
8. J. GURLAND, *Mater. Sci. Eng.* **40** (1979) 73.

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