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Tensile properties of metastable stainless steels

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Tensile properties of two similar stainless steels, types 301 and 304, were investigated for different strain-rates, grain sizes and testing temperatures, and the bulk concentration of martensite formed during deformation was determined by X-ray diffraction.

Results show that type 301 has a distinct strain-peak on the elongation/temperature curve, and strains up to 1.2 were obtained; type 304 steel exhibits only a shallow peak in the elongation/temperature curve. The course of the formation of strain-induced martensite was found to be similar in the two steels at low temperatures (0 to 20°C) but significantly different at higher temperatures. It is shown that it is not the *amount* of martensite that is the major factor in obtaining large elongations to failure, but rather its *formation*, which must be both gradual and selective. It is proposed that the first requirement is dependent on temperature and structural details, while the later requirement is governed by the parameter ($-\ dM/dT$), the temperature sensitivity of the resistance to martensite formation.

1. Introduction

A number of stainless steels are metastable at or near room temperature, i.e. they undergo transformation to martensite on being plastically deformed. Steels of this type include those which contain approximately 18 wt % Cr, 8 wt % Ni, and less than 0.1 wt % C. Normally, these steels have an austenitic structure in the as-received condition, but since the M_d temperature (normally referred to as the temperature of onset of martensitic transformation induced by plastic deformation) is between 0 and 100°C, they transform to martensite when deformed in this temperature range [1-4].

Two direct consequences of this stress-induced transformation are the enhanced strength and the extraordinary tensile elongation of the material [1-6]. Both the mechanism and structural details of the transformation are well understood [7], and the dependence of tensile strength and strain on temperature and strain-rate have also been documented [1, 2, 4, 8].

Mangonon and Thomas have shown that the sequence of reactions in the formation of martensite from austenite is $\gamma \rightarrow \epsilon \rightarrow \alpha$, where ϵ is a hexagonal intermediate phase from which the cubic martensite is nucleated at a later stage; the vol % of ϵ compared to α is negligible above approximately 15% strain. Nucleation of the α martensite was seen to be associated with the intersections of ϵ bands with either twin or grain boundaries. They (and others) have further shown that the yield strength is directly proportional to the amount of α martensite present. Bressanelli and Moskowitz, whose work was mainly restricted to the origin of the large tensile strains in these alloys, stated [4] that the formation of a "specific amount" of martensite during tensile deformation confers resistance to necking.

The role of martensite in promoting very large tensile strains needs some further clarification. The well accepted idea of conferring necking resistance on the material through strengthening by the creation of martensite at incipient necking regions has to be checked in more detail. It seems that the total volume fraction of martensite formed is not the controlling factor, but that the rate of formation and its distribution are also important. The present investigation was undertaken in order to examine the effect of changes of

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temperature, strain-rate, grain size and composition on the course of martensitic transformation during deformation, and hence their overall influence on the tensile properties.

2. Experimental

Two very similar stainless steels were used – types 301 and 304, and their composition is given in Table I. Flat specimens 5 mm wide were prepared from 0.5 mm thick sheet, such that the tensile axes of the specimens were perpendicular to the rolling direction; the effective gauge length for strain measurements was 30 mm. The various heat-treatments and corresponding grain sizes are listed in Table II. All heat-treatments were carried out in an argon atmosphere with the exception of one run where a reduction of the carbon content was deliberately sought. In this case, a batch of specimens of the 301 steel (17DL) was heat-treated at 1150°C in air, resulting in a reduction of the carbon content to 0.04 wt %.

TABLE I

Composition (wt %)	301	304
Chromium	17.7	18.8
Nickel	7.4	9.0
Manganese	1.30	1.15
Carbon	0.08	0.07
Silicon	0.43	—

TABLE II

Specimen	Type	Heat-treatment	Grain size (cm × 10^{-3})
17S	301	600°C, 30 min	3.4
17L	301	1200°C, 30 min	10.6
17DL	301	1150°C, 30 min	5.7
18S	304	600°C, 30 min	2.2
18L	304	1200°C, 60 min	10.8

Testing was carried out on a two-ton Hounsfield Tensometer, specially modified to incorporate a load cell and a controlled temperature bath. The output of the load cell was fed to a Honeywell recorder via a Bruel and Kjaer strain-gauge apparatus. The temperature bath was controlled with a contact mercury controller within ± 0.5°C. Experiments were carried out at two strain-rates, i.e. 6.9×10^{-3} min^{-1} and 6.2×10^{-2} min^{-1}, and at temperatures ranging from 0 to 100°C.

Tensile specimens were examined by X-ray diffraction at an area away from the fracture, using filtered Mo k radiation with a diffractometer scanning speed of $\frac{1}{8}$ degree per minute. The amount of martensite present in the samples was determined from the relative intensities of diffracted beams, following the method of Dickson [9] to correct for modifications to peak intensities caused by the presence of texture. In all, six reflections were examined (three for each phase), and integrated peak intensities determined from the chart record. By comparison with measurements of martensite content from a magnetometer, the accuracy of the X-ray method was found to be ± 1 %; the variation of martensite from specimen to specimen, however, could be as high as ± 5 %.

An attempt was made to follow the course of martensite formation and its distribution by means of metallographic techniques. The results were unsatisfactory, however, because the appearance of the martensite is similar to that of strain markings and therefore the former could be clearly identified only in very lightly deformed specimens. It was possible to identify martensite by transmission electron microscopy, but determination of the overall distribution of martensite by this method is impractical.

3. Results

A set of stress-strain curves (engineering stress and strain) is presented in Fig. 1 for the 17S specimens pulled at a strain-rate of 6.2×10^{-2} min^{-1} and temperatures ranging from 0 to 90°C. Although the curves in the figures have been smoothed out for ease of presentation, it should be noted that they contained many small serrations at high stresses, especially those obtained

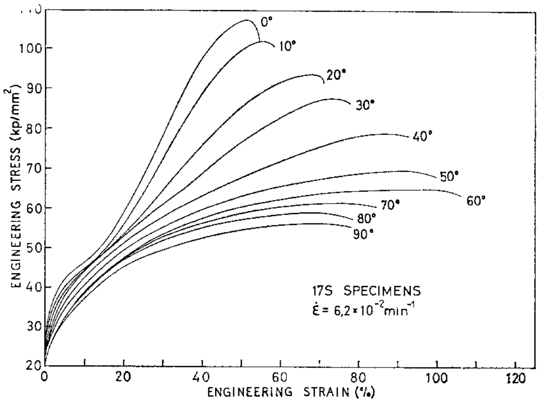

Figure 1 Engineering stress versus engineering strain for the 17S specimens, tested at a strain rate of 6.2×10^{-2} min^{-1}.

at 40, 50 and 60° C. The figures show that both the flow-stress above a strain of about 0.2 and the total strain to fracture are strongly dependent on the testing temperature. The stress-strain curves for all other specimens show a similar dependence on temperature. The marked dependence of flow-stress on temperature at higher strains is due to the occurrence of the martensitic transformation and is well documented. The observation that there is a maximum fracture strain at a particular temperature is the subject of this study.

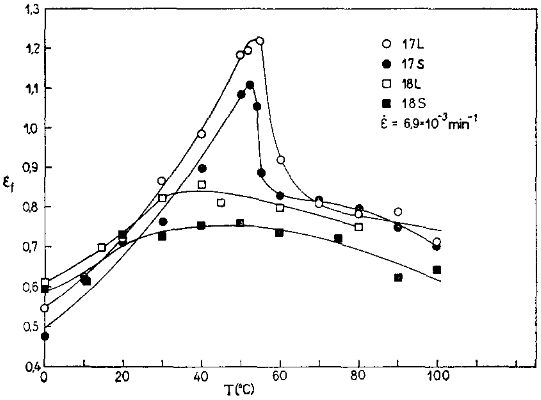

Figure 2 Elongation to failure versus testing temperature for specimens tested at a strain-rate of 6.9×10^{-3} min^{-1}.

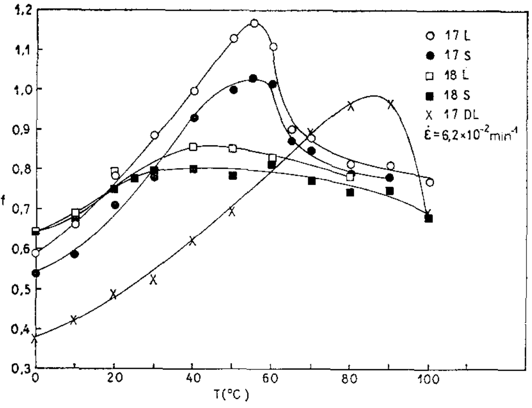

Figure 3 Elongation to failure versus testing temperature for specimens tested at a strain-rate of 6.2×10^{-2} min^{-1}.

Figs. 2 and 3 represent the variation of the total engineering tensile strain to fracture (ϵ_f) with testing temperature. The curves in Fig. 2 correspond to a strain-rate of 6.9×10^{-3} min^{-1}; the upper curves are for the 17/7 steel, and the lower for the 18/8. In Fig. 3, data are plotted for a strain rate of 6.2×10^{-2} min^{-1}, and the curves for the two steels correspond to those in Fig. 2; in addition, the figure contains the plot for the decarburized 17/7 steel. A clear demarcation exists between the three steels; the curves for the 17/7 steels show a fracture strain peak at about 55° C, while that for the decarburized 17/7 is shifted to about 85° C. In contrast to this, the plots for the 18/8 steel have a much more shallow peak, centred approximately on 40 to 45° C. A common feature for all 17/7 specimens (including the decarburized ones) is the asymmetry of the elongation to failure/temperature curves, i.e. a sharp drop in strain values occurs at temperatures above the peak, compared with the more moderate decline below this temperature.

The effect of increasing the grain size of the material is to cause an upward shift of the curves towards greater strain values for both steels. The main effect of decreasing the strain-rate for the 17/7 steel is to produce a sharpening of the peak and slightly greater strain values around the peak temperature. Elongations to failure in excess of 120% were obtained with a large-grained specimen pulled at 55° C at the low strain-rate. With the fine-grained 18/8 steel, the effect of reducing the strain-rate is to lower the curve, in contrast to the behaviour of the 17/7 steel, but there is no significant change with the large-grained 18/8 material when the strain-rate is reduced.

The results of the X-ray determination of the bulk martensite content (as distinct from the martensite content existing in the fracture area) for the various specimens pulled to fracture at different temperatures are presented in Figs. 4 and 5 for the two strain-rates employed. All the curves have a similar form (with the possible exception of the curve for the decarburized 17/7 steel), which is not affected by changes of strain-rate, grain size or composition. The main effect of changing these variables is to produce a shift of the curves. A reduction of the grain-size or strain-rate or an increase in the alloy content, causes a shift of the curve along the temperature axis towards lower temperatures.

Additional experiments were carried out in order to follow the progress of the transformation with increasing strain. Specimens were pulled to various strains at constant temperatures, and the martensite content measured as before; the results are presented in Figs. 6 to 9. For the 18/8 steel at 0° C (Figs. 8 and 9) and for the 17/7 steel at all temperatures (Figs. 6 and 7), the curves have an approximately sigmoidal form. The curves shown in Figs. 6 and 7 and the

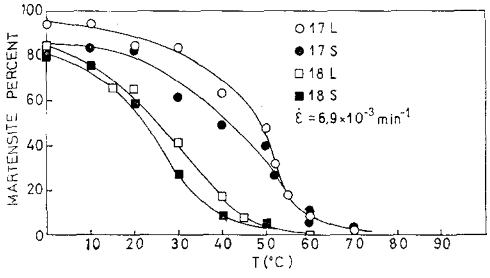

Figure 4 Martensite content of fractured specimens versus testing temperature for a strain-rate of 6.9×10^{-3} min^{-1}.

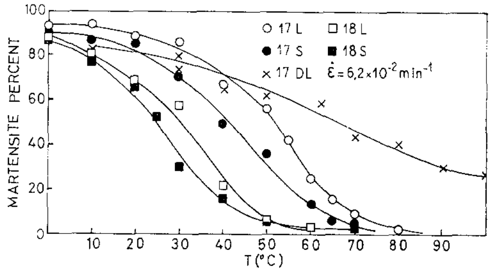

Figure 5 Martensite content of fractured specimens versus testing temperature for a strain-rate of 6.2×10^{-2} min^{-1}.

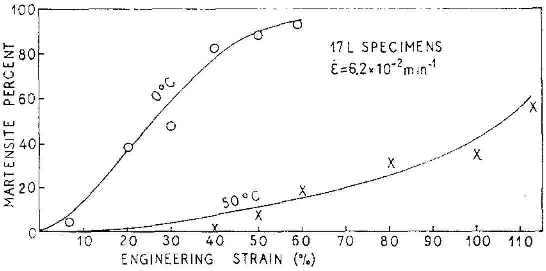

Figure 6 Martensite content versus engineering strain for 17L specimens tested at a strain-rate of 6.2×10^{-2} min^{-1}.

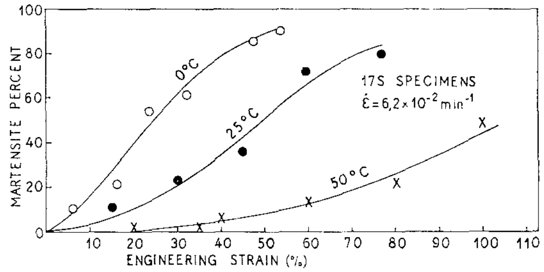

Figure 7 Martensite content versus engineering strain for 17S specimens tested at a strain-rate of 6.2×10^{-2} min^{-1}.

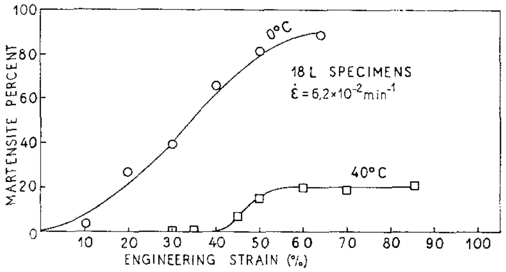

Figure 8 Martensite content versus engineering strain for 18L specimens tested at a strain-rate of 6.2×10^{-2} min^{-1}.

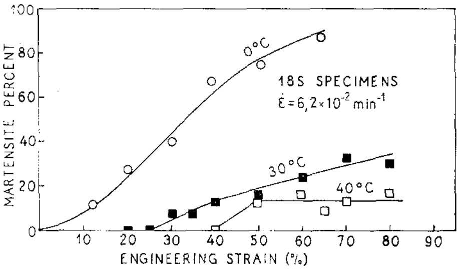

Figure 9 Martensite content versus engineering strain for 18S specimens tested at a strain-rate of 6.2×10^{-2} min^{-1}.

upper curves of Figs. 8 and 9 are sigmoids fitted to the experimental points. With the 18/8 steel tested at 40°C (approximate temperature of the ϵ_f peak), no martensite was observed to form before a certain strain was reached. After this, the martensite content rapidly reaches a constant value. The behaviour of specimens tested between 0 and 40°C is exemplified by the middle curve in Fig. 9, which shows neither the sigmoidal shape of the upper curve, nor the martensite plateau of the lower.

4. Discussion

4.1. Effect of stability on temperature of the ϵ_f peak

Two general trends of behaviour emerge from the results: (1) There is a peak in the ϵ_f/temperature curve for all specimens and the peak temperature is independent of grain size and strain-rate. (2) The composition of the material strongly affects both the sharpness and the position of the peak.

It has already been shown that an elongation to failure peak exists for these types of stainless steels [1, 4]; these peaks are associated with the M_d temperature. Since it is difficult to state M_d with certainty, the temperature at which no more than 5% martensite forms before fracture was arbitrarily fixed as M_d. Accordingly, the M_d temperature for the 17/7 specimens tested at the low and high strain-rates are 65 and 70 to 75°C respectively; for the 18/8 steel, the corresponding temperatures are about 45 to 50°C, and for the decarburized steel, M_d is well in excess of 100°C.

In every case, the temperature corresponding to the ϵ_f peak is lower than the observed M_d, and is unaffected by changes of strain-rate, in contrast to the behaviour of the M_d temperature itself. It is not unexpected that M_d would change with composition, and, because of the observed association between M_d and the peak strain temperature, that the latter would follow the same trend. Of the three alloys, the most stable is the 18/8 steel which has the lowest M_d temperature; the least stable is the decarburized 17/7 steel with the highest M_d, while the remaining steel has an intermediate M_d, and hence, stability.

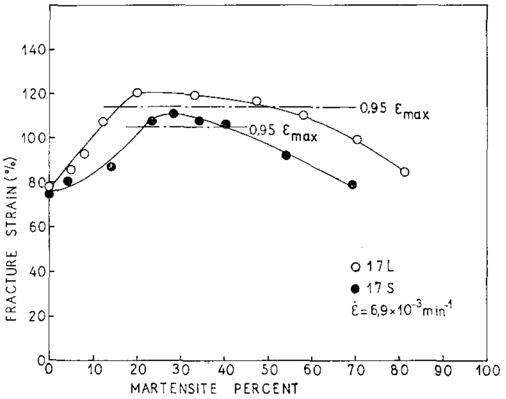

Figure 10 Elongation to failure versus martensite content for 17L and 17S specimens tested at a strain-rate of 6.9 × 10^{-3} min^{-1}.

4.2. Relationship between total martensite content and elongation to failure

The relationship between total martensite content and fracture strain (i.e. elongation to failure) was plotted for two sets of specimens (17L and 17S at 6.9 × 10^{-3} min^{-1}) using data from Figs. 2 and 4. The plot, which is representative for all specimen sets studied, is presented in Fig. 10, where the dashed horizontal lines represent 95% of the peak strain value. For both sets of samples, a large range of martensite contents (15 to 52% for 17L; 20 to 40% for 17S) is associated with elongation to failure values differing by not more than 5% from the peak value. It is significant that the large range of martensite contents (15 to 52%) corresponding to these high strain values for the 17L specimens corresponds to a temperature range of only 9 deg. C. For the 17S specimens, the change in martensite content from 20 to 40%, occurs over only 5 deg. C. Indeed, Figs. 4 and 5 show a maximum slope of the M versus T curves at about the peak strain temperature (55°C). The slope of M versus T curves at any temperature indicate the temperature-sensitivity of martensite formation, and the parameter $-\,\mathrm{d}M/\mathrm{d}T$ is the temperature sensitivity of the resistance to martensite formation.

4.3. Relationship between peak strain temperature and the parameter $-\,\mathrm{d}M/\mathrm{d}T$.

Curves of $-\,\mathrm{d}M/\mathrm{d}T$ for all sets of specimens were obtained by measuring the slopes of all the appropriate curves in Figs. 4 and 5, and presented in Fig. 11. In all cases, the temperature of the maxima coincide reasonably well with the position of the maxima of the ϵ_f versus T curves of Figs. 2 and 3.

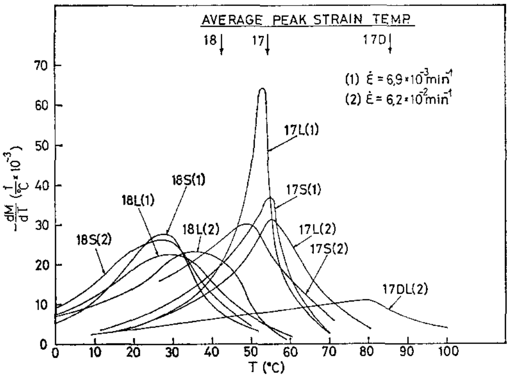

Figure 11 ($-\,\mathrm{d}M/\mathrm{d}T$) versus testing temperature for all specimens studied.

It has already been shown [4] that the role of the transformation in enhancing maximum strains is to confer necking resistance on the material through strengthening by the creation of martensite at preferred sites. Creation of a few martensite plates in an austenite grain in the incipient necking region can already cause hardening of that grain by decreasing the mean free path of the dislocations. This increase in hardening will prevent further deformation there, temporarily preventing further necking at this point. It can be pointed out that the degree of hardening by various degrees of transformation is immaterial. Small and large amounts of martensite in a grain will have the same effect –

stopping further deformation. In order to obtain maximum strain values, the formation of martensite during deformation is required to be both gradual and selective in location. It is proposed that these requirements are met through the controlling action of the parameter $-\mathrm{d}M/\mathrm{d}T$ in the following way. During transformation, localized strain variations within the specimen and associated martensite formation produce corresponding temperature increases [4, 10] in spite of the fact that the specimens are immersed in water during the test. When $-\mathrm{d}M/\mathrm{d}T$ is large, the creation of the martensite in a stressed grain, together with the resultant localized temperature increase, diminish the opportunity for the formation of further martensite at the same site. This temperature-feedback mechanism therefore, results in a uniform distribution of martensite within the strained region. The contradictions between the proposed model and that of Bressanelli and Moskowitz is that they expect a decrease in elongation owing to the local temperature rise, while we suggest an improvement in elongation owing to this temperature rise.

Figs. 2 and 3 show that increasing the grain-size results in higher elongation to failure, especially around the peak temperature. Normally, one would expect the opposite, i.e. specimens with smaller grain size should be more ductile. Grain size has therefore an effect on the gradual and selective distribution of the martensite during the deformation. Further study in this direction is desirable.

4.4 Fracture strain maxima in the 18/8 steel

The dominant strain maxima which characterise the ϵ_f versus T curves for all the 17/7 specimens are not seen in the 18/8 steel, and the very different behaviour of the two sets of steels is quite surprising.

Figs. 8 and 9 show that the course of martensite formation in the 18/8 steel during deformation at 40° C is remarkably different from that of the 18/8 specimens at 0° C, and any of the 17/7 specimens. Here, no martensite forms until a particular strain level is reached, after which the martensite content rapidly reaches a constant value (which persists to fracture) in a relatively short strain interval.

Metallographic examination of specimens of both steels in the unetched condition revealed a high content of finely-dispersed inclusions in the 18/8 steel; no such particles were detected in the 17/7 specimens. It is proposed that during deformation, these second-phase particles set up stress concentrations which in turn trigger the formation of martensite plates uniformly throughout the specimen.

The lack of a peak in the fracture-strain/temperature curves for the 18/8 steel may be attributed, therefore to this non-gradual formation of martensite. At 0° C, the ease of martensite creation outweighs the contribution of the inclusions in nucleating the transformation and hence the M versus ϵ curve has the expected sigmoidal form. At intermediate temperatures (middle curve of Fig. 9), the nucleation contributions are more nearly balanced, and a curve of intermediate form is obtained.

5. Conclusions

1. A definite elongation to failure maximum exists for the 17/7 steels, at a temperature which is unaffected by changes of strain-rate and grain size.
2. A shallow elongation to failure maximum was found for the 18/8 steel, at a temperature which is again unaffected by changes of strain-rate and grain size.
3. For the three steels examined, the temperatures of the strain maxima bear a close relationship to the rate of change of M_d with temperature, i.e. to the degree of stability of the alloys.
4. It is not the *total* amount of martensite, but its distribution which is important in governing the ductility of the materials.
5. The parameter which governs the selectivity of the transformation is considered to be ($-\mathrm{d}M/\mathrm{d}T$) i.e. the temperature-sensitivity of the resistance to martensite formation. In all cases, this term has a maximum value corresponding to the maxima of the fracture strain curves.
6. It is believed that the presence of a dispersed second phase is responsible for the absence of sharp fracture-strain maxima in the 18/8 steel.

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