Effect of plastic deformation on the formation of acicular ferrite

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Abstract

The effect of plastic deformation on the transformation of austenite to acicular ferrite in a Fe–Mn–Si–C alloy steel containing non-metallic inclusions was investigated. The transformation to acicular ferrite is retarded and the final fraction of acicular ferrite is reduced in plastically deformed austenite, which is a characteristic of a displacive transformation mechanism. The increase in the chemical driving force for transformation due to large undercooling below the B_s temperature overcomes the efficacy of dislocations in preventing the growth of acicular ferrite.
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1. Introduction

The recognition and implementation of acicular ferrite in weld metals [1,2] and in wrought steels [3,4] has led to considerable excitement in both industry and academia, primarily because the microstructure offers better toughness. The subject has been reviewed extensively [5–8].

An acicular ferrite microstructure has the potential of combining high strength and high toughness. This is because the plates of acicular ferrite nucleate intragranularly on non-metallic inclusions within large austenite grains, and then radiate in many different orientations from those inclusions whilst maintaining an orientation relationship with the austenite. The resulting microstructure is less organised when compared with ordinary bainite or Widmanstätten ferrite, where there is a tendency to form packets of parallel plates which are identically oriented [5,6]. A crack therefore has to follow a more tortuous path through an acicular ferrite microstructure, thereby leading to an improvement in toughness without compromising strength.

There is evidence to show that in many cases, acicular ferrite is in effect bainite, which nucleates intragranularly on inclusions. For example, the transformation of acicular ferrite occurs below the B_s temperature [9], the transformation exhibits an incomplete-reaction phenomenon [10], the growth of acicular ferrite causes a shape deformation which is characterised as an invariant-plane strain with a large shear component, leading to a stored energy of about 400 J mol^{-1}, which is similar to that of bainite [10,11]. Like bainite, the acicular ferrite always has a well-defined crystallographic orientation relationship with the austenite grain in which it grows, and indeed, is restricted to grow in that grain because the coordinated motion of atoms cannot be sustained across an arbitrary grain boundary [10,11]. There is no partitioning of substitutional alloying elements [11]. Thus, it is expected that the transformation to acicular ferrite, as that of bainite, can be affected by elastic stress [12] and by the plastic deformation of austenite.

When an externally applied stress exceeds the yield strength of austenite, it is possible in principle for the transformation of austenite to bainite to be retarded via a process known as mechanical stabilisation [13–16]. Displacive transformations occur by the advance of glissile interfaces, which can be hindered or rendered sessile on encountering defects such as dislocations or

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grain boundaries. Such defects act as obstacles to the migration of the interface into austenite, rather like the effects which lead to work-hardening when the passage of slip dislocations is obstructed. If the density of dislocations is increased by plastically deforming the austenite, then the dislocations limit the growth of the ferrite plates; the final fraction of bainite may then become smaller in deformed austenite than in undeformed austenite.

Previous work on the mechanical stabilisation of austenite has focused on the martensitic and bainitic transformations. The purpose of the present work was to develop further an understanding of the effect of plastic deformation on the transformation to acicular ferrite in a steel containing non-metallic inclusions, most of which are Mn–Si-oxides.

2. Experimental procedure

2.1. Test material

The alloy composition used in the experiments is Fe–0.22C–2.05Si–3.07Mn–0.7Mo (wt.%). The sulphur and phosphorus concentrations are 0.004 and 0.009 wt.%, respectively; the aluminium, titanium and oxygen concentrations are 0.005, 0.004 and 0.0057 wt.%, respectively.

The alloy contains a relatively high concentration of silicon to retard the precipitation of cementite. The alloy was prepared using high purity materials in a vacuum induction melting furnace. The ingot was forged and machined to a 12-mm diameter rod. Cylindrical samples of 8 mm diameter and 12 mm length were machined from the rods for the Thermecmastor Z (thermomechanical simulator), which can control the temperature of the sample, and the uniaxial load and strain.

2.2. Thermo-mechanical processing

Fig. 1 illustrates the thermo-mechanical cycles used in the experiments. A Pt–Pt/Rh thermocouple was spotwelded half way along the length of the specimen to
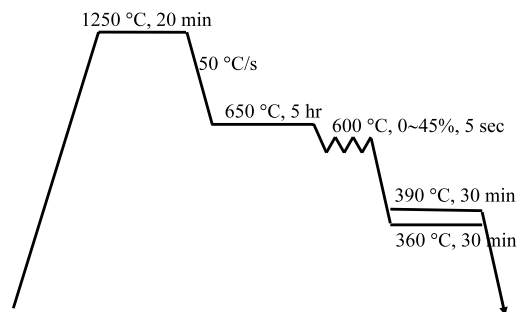

Fig. 1. A schematic diagram showing the thermo-mechanical process used in the present study.

measure the temperature and to provide feedback to the simulator electronics. The cooling between the various isothermal steps was achieved using a computer-controlled helium jet directed at the specimen. The cooling rates were typically 50 °C s^{-1}, except at the final cooling after acicular ferrite transformation. A cooling rate of about 30 °C s^{-1} was achieved at this stage.

Specimens were austenitized at 1250 °C for 20 min to make the austenite grains large in order to encourage the development of acicular ferrite. Specimens were cooled to and maintained at 650 °C for 5 h to form allotriomorphic ferrite at the prior austenite grain boundaries. After cooling to 600 °C, the samples were deformed using a variety of compressive strains in the range 0–45%. Deformed specimens were cooled to 360 or 390 °C and finally isothermally transformed to acicular ferrite at this temperature for 30 min. The volume fraction of acicular ferrite obtained during isothermal heat treatment was evaluated from the radial dilation as measured using a laser extensometer. All specimens for subsequent examination were cut after isothermal heat treatment along the axis of the specimen and the variation of microhardness was measured along the centre-line of each specimen.

2.3. Microscopy

The samples were characterised using optical, scanning electron, and transmission electron microscopy. Samples for optical and scanning electron microscopy were etched using a 2% nital solution. Thin foils for transmission electron microscopy were prepared as 3 mm discs which were then mechanically ground to 100 μm thickness, followed by electro-polishing with 5% perchloric acid and 25% glycerol in methanol solution. The thin foils were examined using a JEOL 200 CX microscope.

3. Results and discussion

3.1. Transformation behaviour

A number of parameters favour the formation of acicular ferrite over ordinary bainite: the presence of non-metallic inclusions, large austenite grain size, and the decoration of austenite grain boundaries with uniform layers of allotriomorphic ferrite. Fig. 2 shows a calculated time–temperature transformation (TTT) diagram for the experimental alloy [17]. Metallography was used to confirm that the heat treatment schedule (Fig. 1), which is based on this calculated TTT diagram, is appropriate to stimulate acicular ferrite. The result is presented in Fig. 3. The mean size of the prior austenite grain measured by line intercept method is 162 μm, its standard deviation is 40 μm, which is larger than the
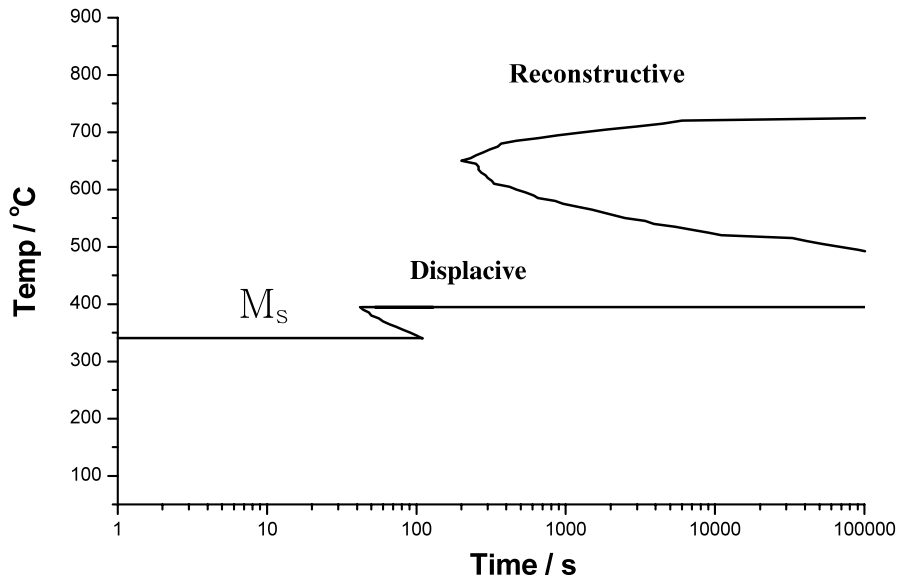
Fig. 2. Calculated TTT diagram of the steel used in the present study, based on model developed by Bhadeshia and [2].

critical size for the transformation of acicular ferrite reported in previous work [9,18–20]. Fig. 3 also shows the allotriomorphic ferrite induced by the isothermal heat treatment at 650 °C for 5 h. The thickness of the

layers of allotriomorphic ferrite is about 2 μm. The volume fraction of allotriomorphic ferrite was estimated to be less than 2%. Acicular ferrite was indeed found after the final isothermal heat treatment.
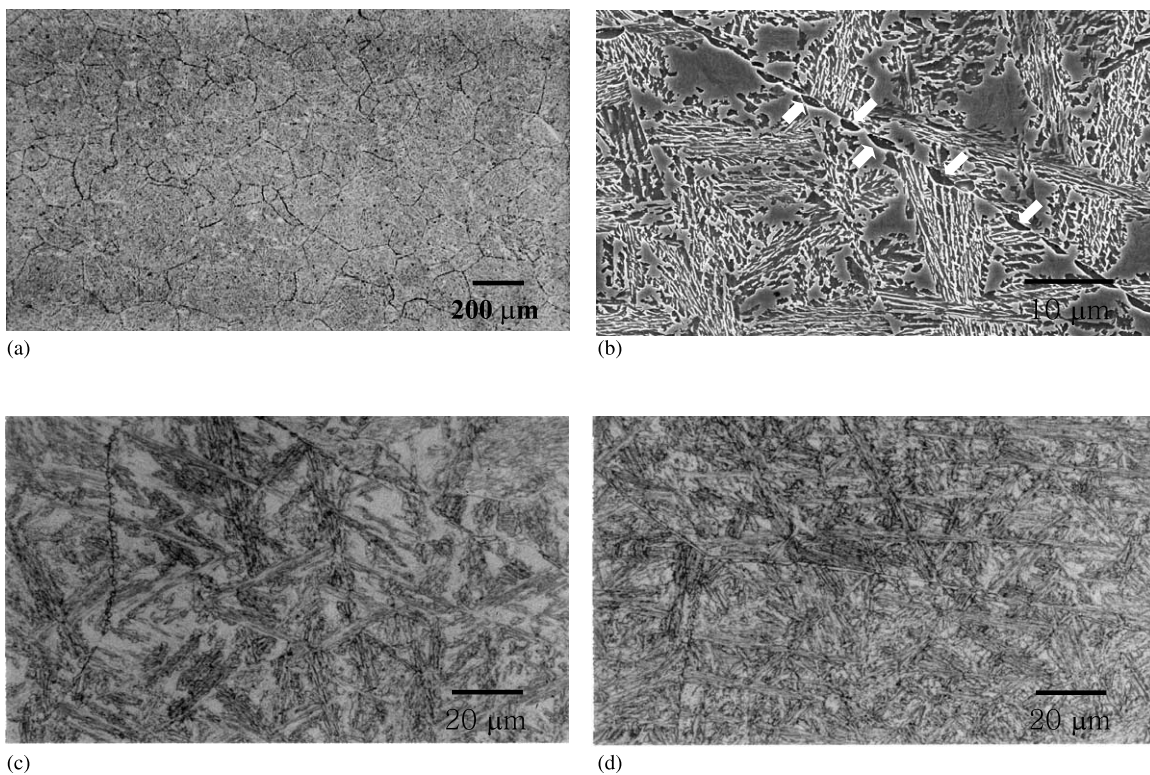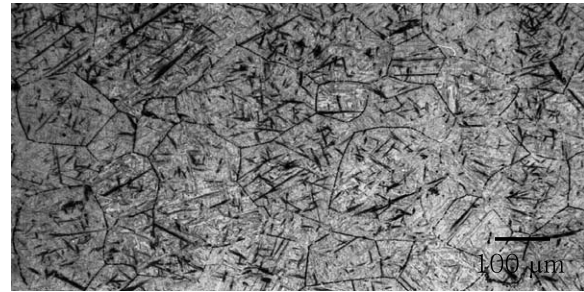

Fig. 3. (a) Optical micrograph showing austenite grain size, (b) scanning electron micrograph showing allotriomorphic ferrite at grain boundaries, (c) optical micrographs showing acicular ferrite transformed at 390 °C in undeformed sample, and (d) acicular ferrite transformed at 360 °C in undeformed sample.

3.2. Effect of allotriomorphic ferrite on acicular ferrite

In general, allotriomorphic ferrite at the prior austenite grain boundaries is helpful to the formation of acicular ferrite because it renders the boundaries inert to the formation of bainite sheaves [5]. An experiment was carried out to confirm this effect. Specimens were austenitized at 1250 °C for 20 min, and then continuously cooled to the isothermal transformation temperature, 390 °C. The microstructure of these samples was compared against others which had been similarly austenitised but then heat-treated at 650 °C for 5 h to form the allotriomorphs, before isothermal transformation at 390 °C. Fig. 4(a) illustrates a typical bainitic microstructure in a specimen which was continuously cooled and then isothermally transformed at 390 °C for 30 min. By contrast, Fig. 4(b) shows acicular ferrite in specimens where a small amount of allotriomorphic ferrite had been introduced prior to transformation below B_s. This result of course supports previous works [9,21] on both weld metal and wrought steel, and provides a convenient method for obtaining acicular ferrite for the purposes of the present work. The partially transformed samples illustrated in Fig. 5 confirm the effect of grain boundary allotriomorphs. The specimen in Fig. 5(a) was quenched after contin-
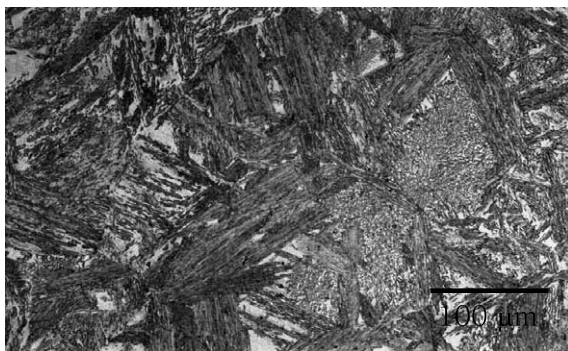

(a)

(b)
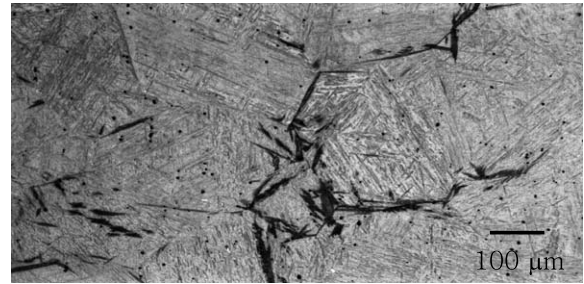
Fig. 5. Optical micrographs showing nucleation at austenite grain boundary in the specimen which was continuously cooled (a) and nucleation within austenite grain in the specimen which was heat treated at 650 °C for 5 h (b)
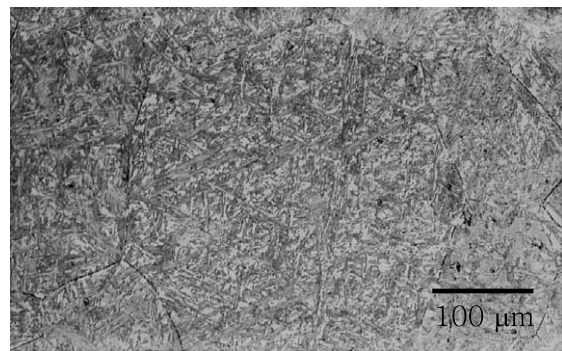
uous cooling and partial transformation at 390 °C for 200 s, and the specimen in Fig. 5(b) was quenched after
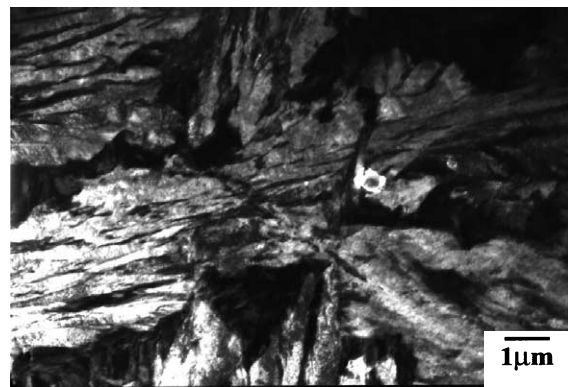

(a)

(b)

(c)

(d)

Fig. 4. Microstructures from the sample which was continuously cooled and followed by isothermal transformation at 390 °C for 30 min (a and c) and the sample which was heat treated 650 °C for 5 h and followed by isothermal transformation at 390 °C for 30 min (b and d); (a and b) optical micrographs, (c and d) transmission electron micrographs
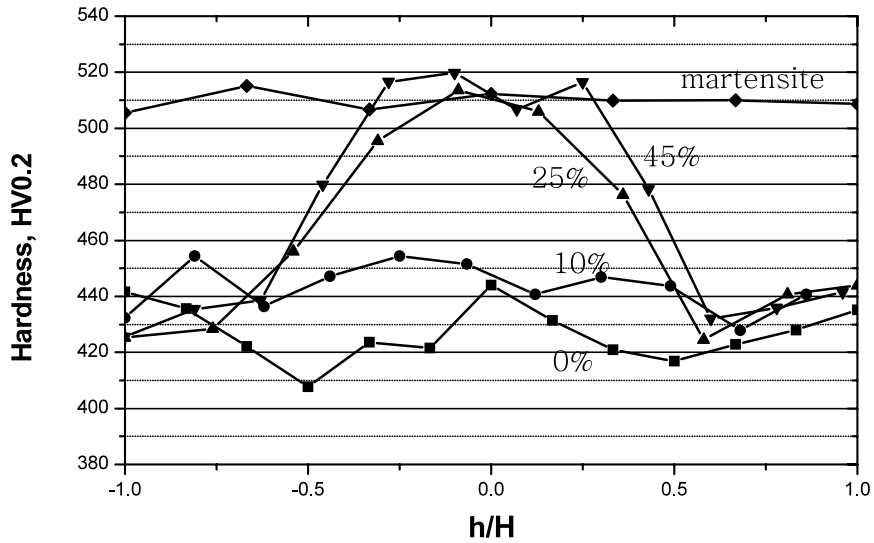
Fig. 6. Hardness profiles along the centreline of specimens isothermally transformed at 390 °C after deformation with various amounts of strain (*H* is cylinder half height, *h* is distance from centre of specimen).

isothermal heat treatment at 600 °C for 5 h and partial transformation at 390 °C for 100 s. In Fig. 5(b), where grain boundaries are delineated with allotriomorphic ferrite, intragranular acicular ferrites are well developed.

3.3. Effect of total strain

The degree of deformation (and hence the extent of mechanical stabilisation of austenite) in the samples studied using the thermo-mechanical simulator is not homogeneous. As is well known, the ends of specimens are constrained by the compression anvils and hence deformed less than the central regions, an effect often described as barrelling. This phenomenon has been used

to advantage in the present work to study the effect of different strains present in individual specimens, on the progress of transformation.

Fig. 6 shows the variation of hardness along the centre-line of the specimens after isothermal transformation at 390 °C as a function of the amount of deformation. The hardnesses of acicular ferrite and martensite in undeformed specimens showed uniform values of about 430 HV0.2 and 510 HV0.2, respectively.

However, the specimens deformed prior to transformation exhibit significant hardness variations along the centre-line. As expected, the deformed samples are harder at the centre when compared with those which are undeformed. Their hardness values approach the
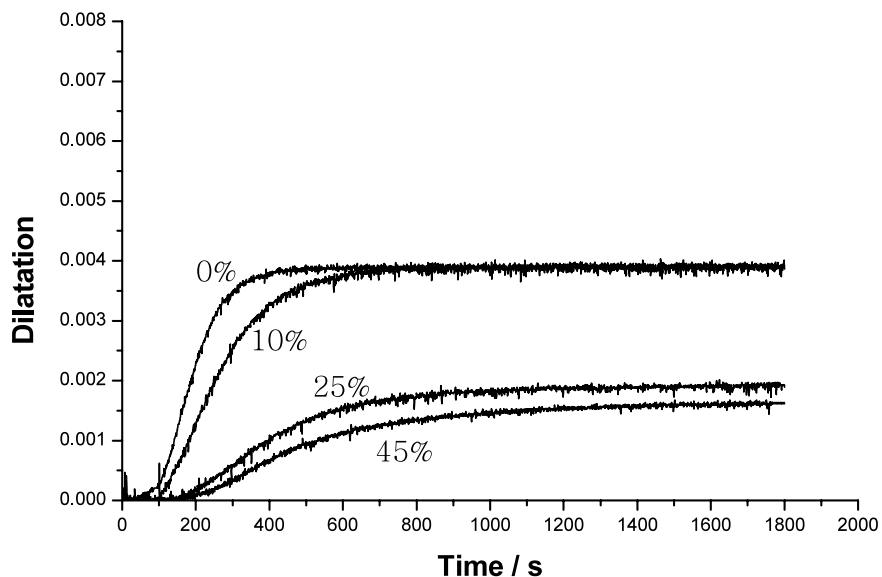

Fig. 7. Radial dilatation change during transformation to acicular ferrite/bainite at 390 °C after various amounts of deformation.
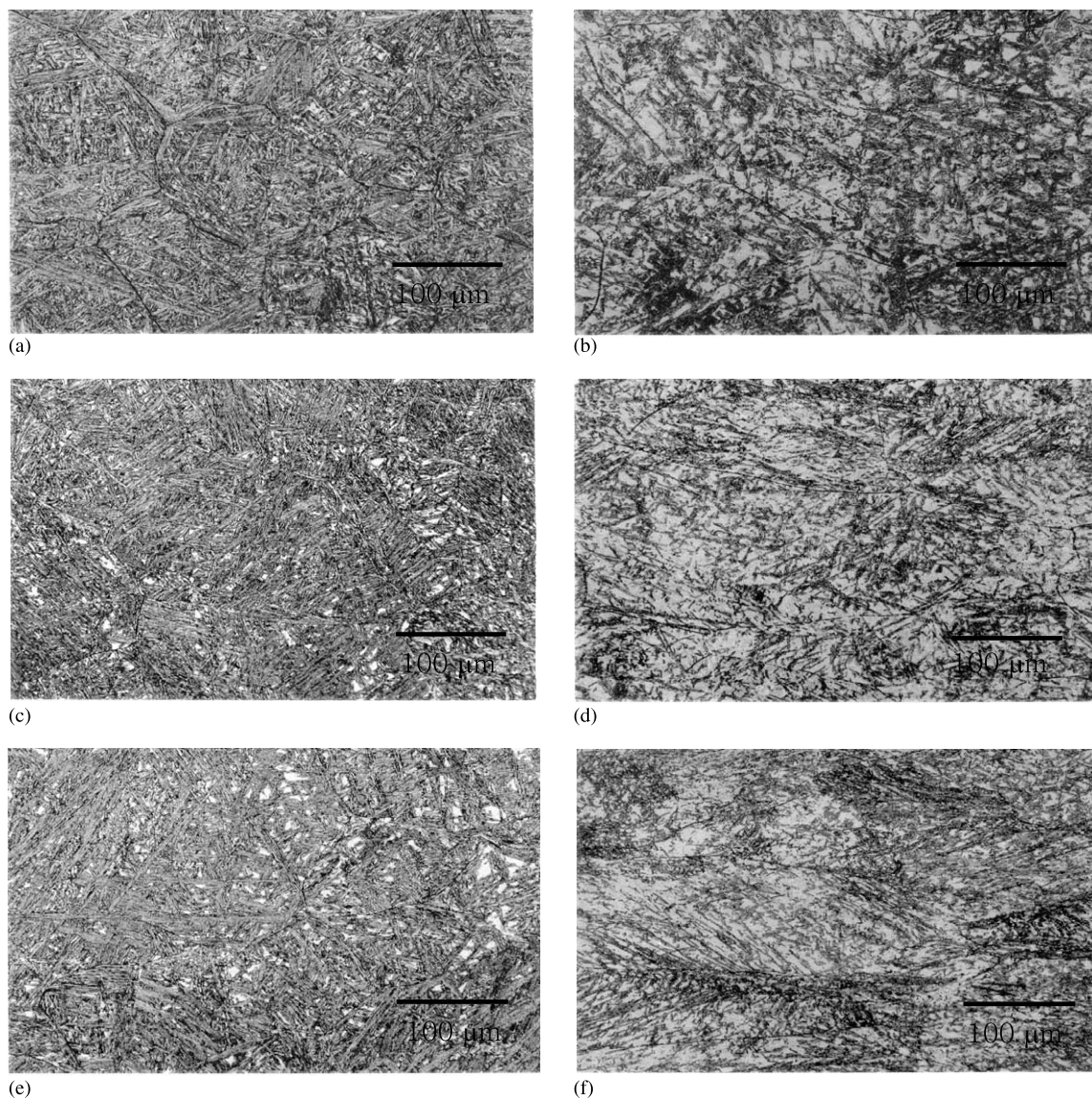
Fig. 8. Comparison of microstructure at the end and centre of specimens isothermally transformed at 390 °C after deformation: (a) the end of 10%-deformed sample (b) the centre of 10%-deformed sample (c) the end of 25%-deformed sample (d) the centre of 25%-deformed sample (e) the end of 45%-deformed sample (f) the centre of 45%-deformed sample.

value of a fully martensitic microstructure. On the other hand, the hardness values at the ends of the deformed specimens are quite comparable to those of specimens in which the austenite is not deformed prior to transformation. The fraction of acicular ferrites becomes smaller, and that of martensite becomes larger, in the regions of the sample which are most severely deformed, proving that austenite can be mechanically stabilised to the formation of acicular ferrite.

This result was confirmed by dilatometric measurements (Fig. 7) and also using metallography (Fig. 8). Fig. 7 shows that the measured radial decreases with increasing prior deformation, especially when the strain is 25% or higher. The final fraction of transformation in the 25% deformed specimen was much less than in the sample with only 10% prior strain. Fig. 8 shows that

almost complete transformation is achieved at the undeformed-ends of the specimen with only partial transformation at the centre. The ferrite plates are also refined with increasing deformation.

3.4. Effect of transformation temperature

The effect of isothermal transformation temperature was investigated by lowering the temperature to 360 °C without changing other thermo-mechanical treatment conditions. Fig. 9 shows the resulting hardness profiles. The hardness values at the centre and at the end of all compressed specimens are almost identical to those of an undeformed specimen. In other words, the degree of mechanical stabilisation is reduced at large undercoolings. This is expected since the driving force for
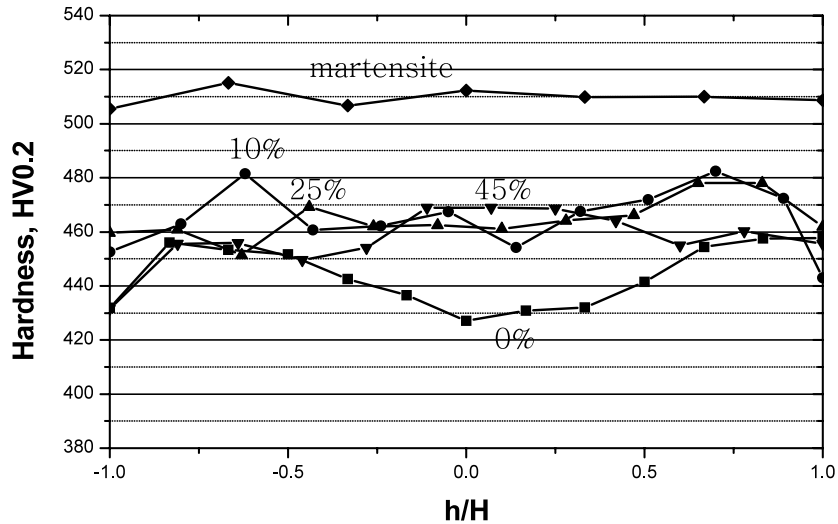
Fig. 9. Hardness profiles along the centreline of specimens isothermally transformed at 360 °C after deformation with various amounts of strain (*H* is cylinder half height, *h* is distance from centre of specimen).
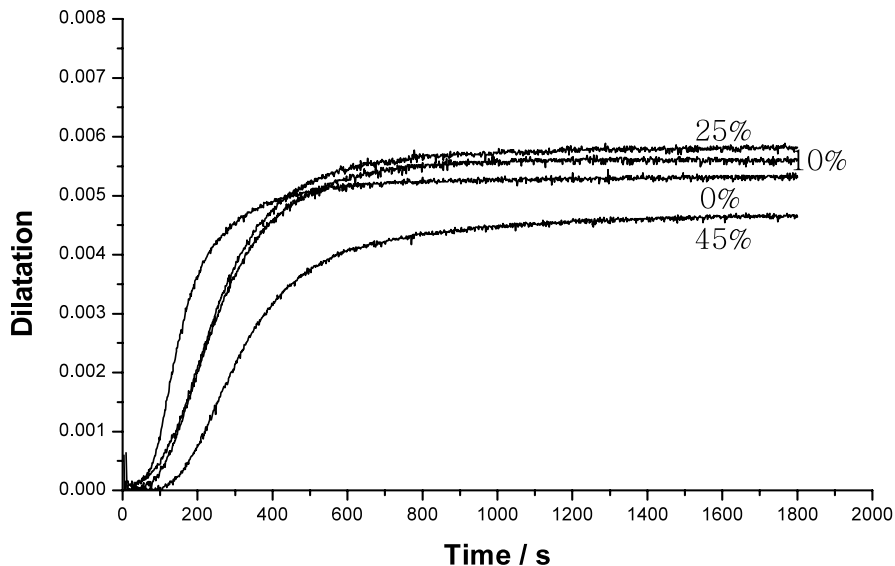

Fig. 10. Radial dilatation change during transformation to acicular ferrite/bainite at 360 °C after various amounts of deformation.

transformation becomes larger. The dilatometric results (Fig. 10) and the optical micrographs (Fig. 11) support these observations. A small degree of mechanical stabilisation was evident in the samples strained to 45%, the largest deformation used. The prior deformation nevertheless led to a reduction in the size of the acicular ferrite plates, probably because of its effect on enhancing the nucleation rate of acicular ferrite.

4. Conclusions

Consistent with published work, it has been confirmed that the presence of allotriomorphic ferrite at the austenite grain boundaries renders them ineffective for the nucleation of bainite, even though the number density of grain boundary nucleation sites is increased by deformation. Consistent with the known mechanism of transformation it has been demonstrated that the growth of acicular ferrite is retarded in plastically deformed austenite. This is a key characteristic of a displacive transformation mechanism. Furthermore, the impediments to growth of acicular ferrite in deformed austenite have, as expected, shown to be less effective as the chemical driving force for transformation is increased at large undercoolings below the B_s temperature.
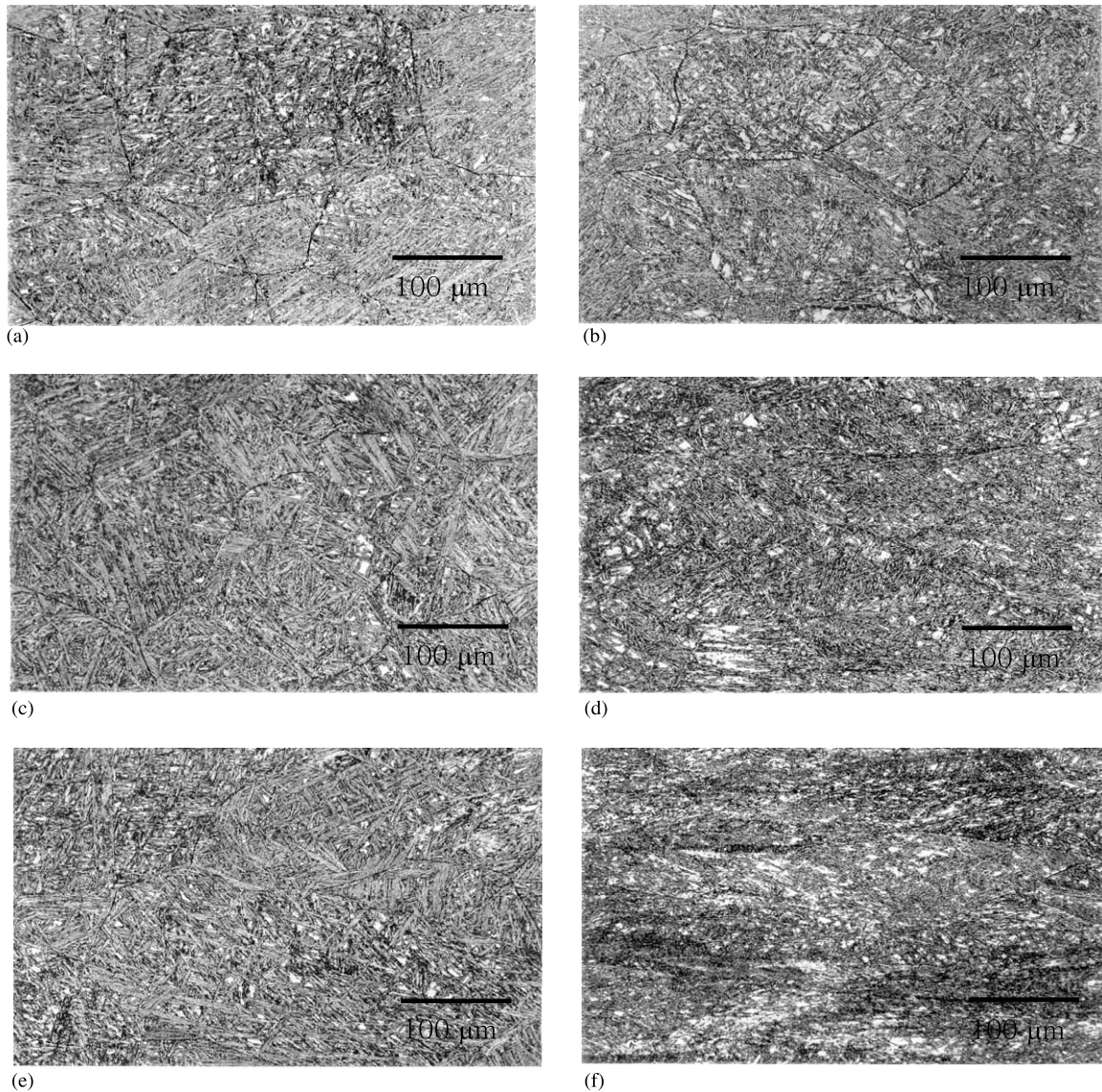
Fig. 11. Comparison of microstructure at the end and centre of specimens isothermally transformed at 360 °C after deformation: (a) the end of 10%-deformed sample (b) the centre of 10%-deformed sample (c) the end of 25%-deformed sample (d) the centre of 25%-deformed sample (e) the end of 45%-deformed sample (f) the centre of 45%-deformed sample.

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