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DEFORMATION AND RECRYSTALLIZATION TEXTURES IN TWO DUAL PHASE STEELS

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Abstract

Dual phase microstructures were produced in a C-Si-V and C-Mn-V steel by intercritical annealing of initial martensitic structures, followed by water quenching. Both the steels were cold deformed 60% at room temperature and then subjected to recrystallization anneals at 650°C and 800°C. The major texture components in the cold deformed alloys are {111} <112> and {111} <110>. The textures of the recrystallized alloys also contain basically the same texture components. A reasonably strong and perfect <111>||ND fibre texture is observed in the C-Mn-V steel, whereas this fibre is weaker and rather imperfect in the C-Si-V steel. These textural differences are reflected in the $\bar{r}$-values obtained for the two steels.

Introduction

Dual phase steels show wide veriety of textures consequent on recrystallization [1-3]. It is known that the {111}- type of fibre texture is favourable for deep drawing of sheet steels. Therefore it would be desirable to develop the {111} fibre texture in dual phase steels by suitable processing. With this view in mind the present work was undertaken to examine the deformation and recrystallization textures in two different dual phase steel compositions. The texture results were correlated with corresponding $\bar{r}$-values.

Experimental procedure

The two steels, A1: 0.11C, 1.50Si, 0.09V, 0.011S, 0.018P (wt. percent) and A3: 0.11C, 1.48Mn,.08V, 0.027S, 0.014P (wt.percent), were hot forged, solution treated at 910°C and quenched in water to produce an initial martensitic structure (designated as WQ). Specimens from the quenched material were subjected to intercritical heat treatment at 750°C and 810°C for 30 minutes followed by water quenching. Structures produced by the above treatments are designated as WQ750/WQ and WQ810/WQ,respectively. The intercritically annealed materials were cold rolled to a reduction of 60% in thickness. Samples cut from each cold rolled sheet were recrystallization annealed for various lengths of time in salt baths kept at 650°C and 800°C.

Crystallographic textures from the cold rolled and the recrystallized samples were determined by measuring {110}, {200} and {112} pole-figures in an automatic texture goniometer [4]. Three dimensional orientation distribution functions (O.D.F's) were calculated from these pole-figure data in the usual manner [5,6]. In the O.D.F. method the orientation of a crystallite is specified with respect to the specimen co-ordinate system by the three Eulerian angles ψ_1, ϕ and ψ_2 and an orientation distribution function, f(g), representing the relative intensity of orientations, is computed. The Lankford parameter, r, was calculated theoretically on the basis of {hkl} <111> pencil glide from the experimentally determined O.D.F's [7]. The plastic strain ratio ($\bar{r}$) was then determined on the basis of the formula $\bar{r}=(r_0+2r_{45}+r_{90})/4$, where the numerical subscripts denote the angles from the rolling direction.

Results and discussion

Deformation textures

Typical O.D.F. plots (Fig. 1a and b), obtained from the cold rolled material, show sharp deformation textures of the type {111} <110> and {111} <112> along with some minor components, namely, {337} <110>, {337} <776>, {112} <111> and {112} < 110> in both the alloys A1 and A3. The texture is much sharper in alloy A1 as compared to alloy A3. Incidentally, the alloy A1 contains less martensite (20% for WQ750/WQ and 24% for WQ810/WQ), whereas alloy A3 possesses much higher amounts of martensite (28% for WQ750/ WQ and 32% for WQ 810/WQ). Martensite, being the harder phase, its presence in the microstructure is expected to inhibit easy plastic flow in the ferrite matrix, thus making re-orientation of the grains more difficult in order to produce sharp deformation texture. The sharper texture obtained in either of the cold rolled alloys where the dual phase structure was produced by intercritical annealing at the lower temperature of 750°C as compared to 810°C, can also be related to the lesser amount of martensite in the former as compared to the latter. The development of {111} fibre texture in the cold rolled alloys was checked by plotting f(g)-values against ψ_1 for ϕ =55° and ψ_2=45°. The above plots for both the alloys are given in Figure 2. Evidently, the individual components in the {111} <uvw> textures do not develop into a strong fibre in alloy A1, whereas a reasonably strong <111>||ND fibre texture is noticed in the cold rolled alloy A3.

Recrystallization textures

O.D.F's obtained from cold rolled and recrystallized samples of both the alloys indicate that the deformation and recrystallization textures are rather similar. Typical O.D.F. plots of textures for the recrystallized alloys are shown in Figures 3 (a-d) for the material prepared by intercritical annealing at 750°C before cold rolling. For the different recrystallized samples of both the alloys the f(g)-values from the O.D.F's were plotted against ψ_1 for ϕ =55° and ψ_2=45° and these are shown in Figures 4 and 5. It is clearly seen from these figures that whereas the {111} <uvw> orientations do not constitute a complete fibre in alloy A1, alloy A3 exhibits a sharp {111} fibre. The intensities of the {111} <uvw> components are stronger in both the alloys after recrystallization at the higher temperature of 800°C, as compared to 650°C. Finely dispersed particles (unidentified) are frequently observed in both the alloys, recrystallized at the lower temperature of 650°C (Fig. 6a). Such particles are normally believed to retard the recrystallization kinetics, having a greater influence on nucleation than on growth [8]. In the absence of such particles at the higher temperature of recrystallization of 800°C (Fig.6b), it is expected that more number of {111} -nuclei will have an easy and uninterrupted growth giving rise to a sharper {111} texture.

$\bar{r}$-values

A look at the $\bar{r}$-values of the recrystallized alloys [Table 1] reveals that alloy A1 always has $\bar{r} < 1$, whereas alloy A3 has $\bar{r} > 1$. Thus alloy A3 seems to have some advantage over alloy A1 in terms of deep-drawability. It may be recalled that, the sharpest value for the intensity of the <111>||ND fibre texture has been obtained for alloy A3 recrystallized at 800°C. This suggests that sharper the {111} fibre texture larger is the $\bar{r}$-value. In addition to this, the somewhat higher $\bar{r}$-values obtained for the samples, initially dual phase treated at 750°C, also indicate that the lower initial martensite content in the steel is beneficial in enhancing the $\bar{r}$-values. However, the generally rather poor $\bar{r}$-values obtained for the two experimental steels clearly points to the desirability to explore means other than simple cold rolling and recrystallization in order to improve the deep-drawability of such steels.

Table 1 $\bar{r}$ -values of recrystallized alloys

Alloy designation and Heat treatment	$\bar{r}$-values Recrystallized at 650°C	$\bar{r}$-values Recrystallized at 800°C
A1/WQ750/WQ	0.84	0.96
A1/WQ810/WQ	0.71	0.82
A3/WQ750/WQ	1.02	1.19
A3/WQ810/WQ	1.05	1.10

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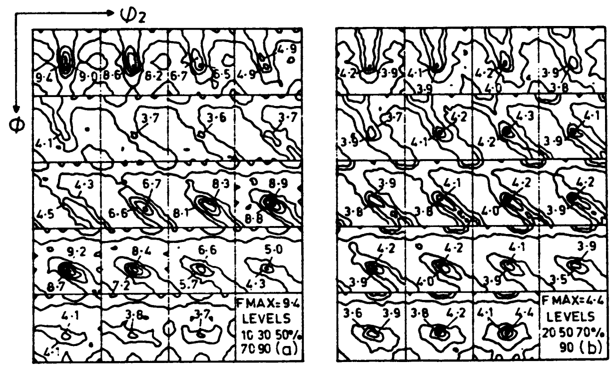

Fig.1. O.D.F's showing φ_1 sections for
(a) A1/WQ750/WQ cold rolled 60%
(b) A3/WQ750/WQ cold rolled 60%

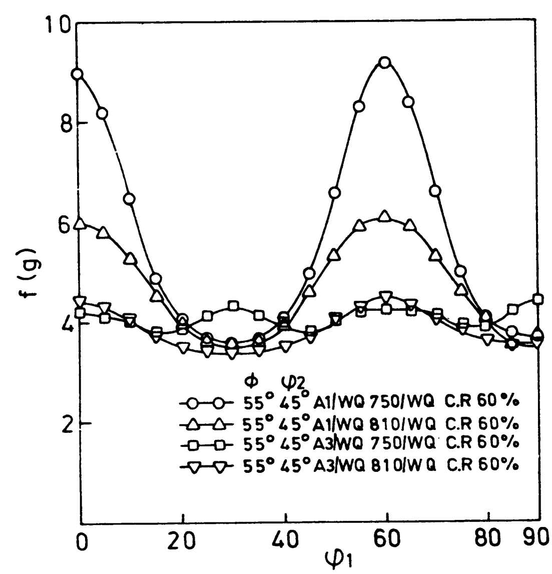

Fig.2 Variation of f(g) with φ_1 along Φ/φ_2 line for the cold rolled alloys A1 and A3

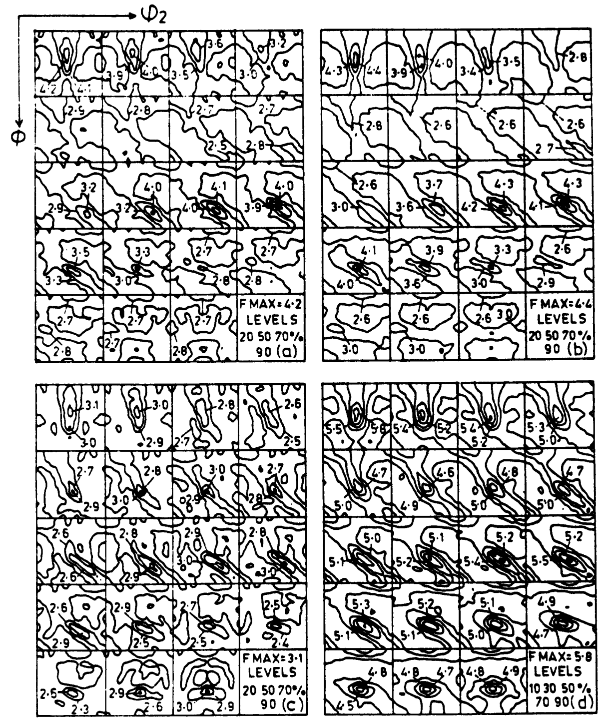

Fig. 3. O.D.F's showing φ_1 sections for
A1/WQ750/WQ: (a) Recryst. at 650°C
(b) Recryst. at 800°C
A3/WQ750/WQ: (c) Recryst. at 650°C
(d) Recryst. at 800°C

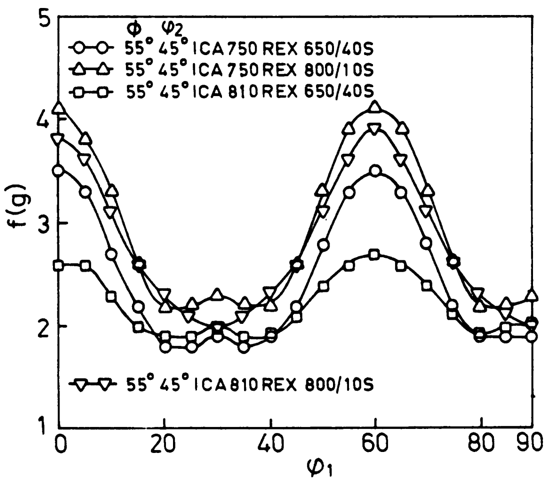

Fig. 4. Variation of f(g) with φ_1 along Φ/φ_2 line for the recrystallized alloy A1

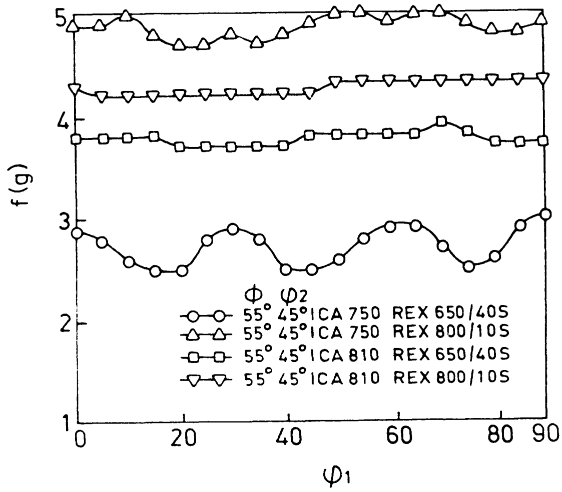

Fig. 5. Variation of f(g) with φ_1 along Φ/φ_2 line for the recrystallized alloy A3

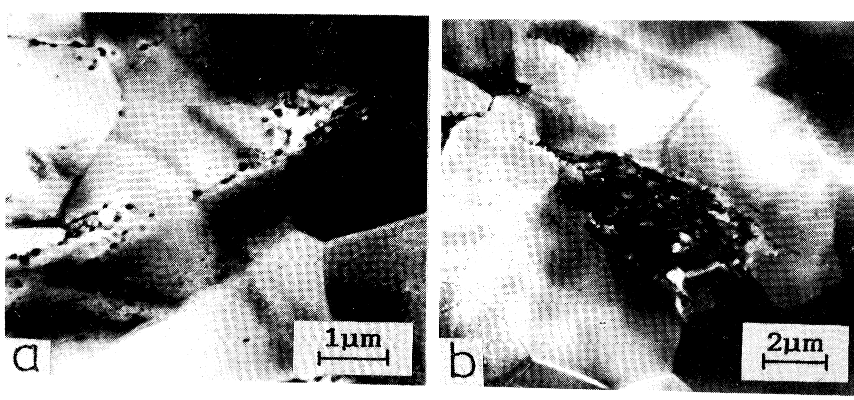

Fig. 6 TEM photographs of the alloy Al
(a) recrystallized at 650°C
(b) recrystallized at 800°C