Development of Primary Recrystallization Textures in Metastable Austenitic Steels

BY M.J. DICKSON AND R.P. STRATTON
Sheffield Polytechnic, Pond Street, Sheffield, S1 1WB, England

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The cold-rolling and annealing textures of 18%Cr-10%Ni, 18%Cr-12%Ni and 18%Cr-14%Ni steels have been examined with particular reference to the role of the martensite transformation in developing primary recrystallization textures. The amount of martensite formed after 90-95% cold rolling increases from ~5% for 18%Cr-14%Ni steel to ~86% for 18%Cr-10%Ni steel, and textures of both α' martensite and the remaining austenite are the same for each steel. Principal components of the martensite texture are {112} (T10) and {111} (T2), with a minor component {001} (T10). The austenite exhibits a normal f.c.c. alloy type texture, {110} (T12) + a minor {110} (001) component. Both 18%Cr-12%Ni and 18%Cr-14%Ni steels recrystallize to a {225} (3-33) texture, whereas the 18%Cr-10%Ni steel recrystallizes to mainly {230} (3,2,14) and {6,10,1} (531) orientations with considerable spread remaining in the region of {6,8,17} (575), which is close to {225} (575).

Factors influencing development of the annealing textures are the relative amounts of α' and γ formed by cold rolling, recovery in both phases, γ → α' shear transformation, variant selectivity during the transformation and, finally, competitive growth of favourably oriented nuclei.

Experimental

The materials were supplied in the form of 1/4 in forged slabs which had been annealed for 1 hour at 1050°C. Compositions are given in Table 1.

Preliminary working consisted of cold rolling ≥30% followed by annealing for 1 hour at 1050°C, the sequence being repeated down to a thickness of 0.2 in. The alloys were pickled in a boiling solution of 5% nitric acid, 30% hydrochloric acid, 65% water after each heat treatment.

Samples representing this condition were retained for texture examination. These were machined on one side only to a thickness of 0.15 in, followed by grinding and etching. Diffractometer patterns were obtained using Mo Kα radiation and texture parameters, ϑhkl, were calculated to define this effective starting condition. Definition and significance of these parameters have been discussed previously (Dickson, 1969).

In a previous examination of 18%Cr steels (Dickson & Green, 1969) it was shown that complete stability occurs with nickel contents > ~14%, e.g. 18%Cr-14%Ni steel (γ/Γb = 2.09 x 10^-3) contains ~5%α' after cold rolling and represents one extreme of a range of compositions which give rise to duplex textures during rolling. The opposite extreme is typified by the behaviour of 18%Cr-10%Ni steel (γ/Γb = 1.43 x 10^-3) which transforms to ~86% martensite.

The purpose here is to reconsider, in the light of further experimental work, an analysis of textures in these alloys, with particular reference to the role of the martensite transformation and the overall development of primary recrystallization textures.
RECRYSTALLIZATION TEXTURES IN METASTABLE AUSTENITIC STEELS.

Table 1. Compositions of Steels (wt. %)

<table>
<thead>
<tr>
<th>Steel</th>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>S</th>
<th>P</th>
<th>Cr</th>
<th>Ni</th>
<th>N</th>
<th>Fe</th>
</tr>
</thead>
<tbody>
<tr>
<td>18/10</td>
<td>0.026</td>
<td>0.1</td>
<td>0.2</td>
<td>0.005</td>
<td>0.005</td>
<td>18.2</td>
<td>10.0</td>
<td>0.012</td>
<td>Balance</td>
</tr>
<tr>
<td>18/12</td>
<td>0.032</td>
<td>0.2</td>
<td>0.2</td>
<td>0.006</td>
<td>0.005</td>
<td>18.0</td>
<td>11.9</td>
<td>0.010</td>
<td>Balance</td>
</tr>
<tr>
<td>18/14</td>
<td>0.017</td>
<td>0.1</td>
<td>0.2</td>
<td>0.005</td>
<td>0.005</td>
<td>18.3</td>
<td>14.0</td>
<td>0.011</td>
<td>Balance</td>
</tr>
</tbody>
</table>

changes during annealing, intensities of particular components have been assessed from point measurements taken from pole figure data.

Phase analysis of the cold-rolled and annealed specimens was by a method described previously (Dickson, 1969). Lattice re-arrangement during the early stages of annealing was studied from line profiles of selected Bragg reflexions.

Results

Table 2 shows that each steel had an essentially random grain orientation prior to final cold rolling, so that subsequent influence of initial texture as a variable can be dismissed.

Table 2. Texture parameters after preliminary rolling and annealing

<table>
<thead>
<tr>
<th>Steel</th>
<th>P_{hkl}</th>
<th>200</th>
<th>220</th>
<th>311</th>
<th>331</th>
<th>420</th>
<th>422</th>
</tr>
</thead>
<tbody>
<tr>
<td>18/10</td>
<td>1.1</td>
<td>0.8</td>
<td>1.2</td>
<td>1.1</td>
<td>0.9</td>
<td>1.0</td>
<td>0.9</td>
</tr>
<tr>
<td>18/12</td>
<td>1.2</td>
<td>0.9</td>
<td>1.2</td>
<td>0.9</td>
<td>1.1</td>
<td>0.8</td>
<td>0.9</td>
</tr>
<tr>
<td>18/14</td>
<td>1.1</td>
<td>0.9</td>
<td>1.1</td>
<td>1.1</td>
<td>0.8</td>
<td>1.0</td>
<td>1.0</td>
</tr>
</tbody>
</table>

Fig. 1 shows the progress of martensite formation during rolling. All steels exhibit an 'incubation period' during which the $M_\alpha$ is raised to room temperature. Maximum amount of martensite is formed by $\sim 90\%$ reduction.

A selection of {200} pole figures are shown for the $18\%$Cr--$10\%$Ni and $18\%$Cr--$14\%$Ni steels after $95\%$ cold rolling. The intensity contours are labelled in multiples of the random level. Textures were the same after $90\%$ cold rolling, with intensities slightly lower. Pole figures of $18\%$Cr--$12\%$Ni steel have been omitted since they were similar to those of $18\%$Cr--$14\%$Ni steel.

$18\%$Cr--$10\%$Ni steel

The $\alpha'$ texture of $18\%$Cr--$10\%$Ni steel [Fig. 2(a)] really consists of two fibre components, i.e. a $\langle 111 \rangle$ fibre component normal to the rolling plane and an incomplete $\langle 110 \rangle$ fibre texture parallel to the rolling direction. The principal orientations are $\{ 112 \} \langle 110 \rangle$ and $\{ 111 \} \langle 112 \rangle$ with spread towards $\{ 001 \} \langle 110 \rangle$. The austenite exhibits a normal f.c.c. alloy type of texture, viz. $\{ 110 \} \langle 112 \rangle$ + a minor $\{ 110 \} \langle 001 \rangle$ component [Fig. 2(b)].

Fig. 3 compares relative intensities of the principal $\alpha'$ components and Fig. 4 shows profiles of the 200 $\alpha'$ reflexion. These measurements have been made using identical instrument settings, so that they are a function not only of textural changes but also of the amount of martensite present. The intensity of all components increases after $\frac{1}{2}$ hour at $500^\circ$C, when the amount of martensite has actually decreased slightly. This is related to the onset of recovery, causing sharpening of the $\alpha'$ texture. There is no evidence of recrystallization within the martensite at $600^\circ$C, and at $700^\circ$C the amount of martensite is so small that recrystallization cannot be entirely eliminated but the breadth of the line does suggest that the amount of recrystallization, if any, is minimal.

The $y$ pole figure after annealing for $\frac{1}{2}$ hour at $500^\circ$C is not shown, but the texture at this stage was similar although slightly less well defined than at $600^\circ$C. After annealing for $\frac{1}{2}$ hour at $600^\circ$C only $\sim 10\%$ $\alpha'$ remains (Fig. 3) and one of the peak distributions is centred on $\{ 110 \} \langle 001 \rangle$ with spread towards $\{ 110 \} \langle 112 \rangle$ [Fig. 5(a)]. For reasons which will become evident during the discussion, the orientations $\{ 110 \} \langle 557 \rangle$, $\{ 430 \} \langle 001 \rangle$ and $\{ 430 \} \langle 6, 8, 17 \rangle$ are also shown as belonging to this part of the texture. The other high-intensity region in Fig. 5(a) is centred on orientations close to $\{ 6, 8, 17 \} \langle 575 \rangle$ but includes considerable spread.

Fig. 1. Variation of $\% \alpha'$ during cold rolling of $18\%$Cr steels containing 10, 12 and 14$\%$ Ni.
Fig. 2. Pole figures of 18% Cr–10% Ni steel cold rolled to 95% reduction: (a) [200]α' pole figure, (b) [200]γ pole figure.

towards {430}(340). Again, this will later be shown to be quite significant. At 700°C the martensite has almost disappeared (Fig. 3) and Fig. 5(b) shows that, in the region {110}⟨112⟩–{110}⟨001⟩, the distribution of intensity has changed and the general level of intensity has decreased. At the same time, in the region of {430}⟨340⟩, the intensity has increased.

The positions of {230} ⟨3,2,14⟩ and {6,10,1} ⟨531⟩ orientations are also indicated in Fig. 5(b), to show that they lie within the general spread of texture at this stage. The intensity of these components increases after annealing at 800°C, and after ½ hour at 900°C [Fig. 5(c)] they emerge as major components of the recrystallization texture. There still remains a considerable spread towards other orientations, particularly those of the type {6,8,17} ⟨375⟩.

Fig. 6 compares intensities of important γ orientations and Fig. 7 shows profiles of the 220γ reflexion. The slight increase in line intensity at 500°C (Fig. 7) could be partly due to a decrease in the amount of martensite. However, having identified recovery within the martensite at 500°C, it is reasonable to assume that it has also started in the austenite at this temperature. (This is further confirmed when the results for 18%Cr–12%Ni and 18%Cr–14%Ni steels are considered). Recovery within the austenite is certainly evident at 600°C. Returning to Fig. 6, the initial increase in intensity of {110} ⟨001⟩ and {110} ⟨112⟩ coincides with the transformation of α' → γ, with recovery taking place simultaneously in both phases. The intensities of {6,10,1} ⟨531⟩, {6,8,17} ⟨375⟩ and {230} ⟨3,2,14⟩ components increase during the recovery range and continue to increase up to 900°C. The intensities of {110} ⟨112⟩ and {110} ⟨001⟩ decrease during recrystallization.

18%Cr–12%Ni and 18%Cr–14%Ni steels

Rolling textures remain the same as the nickel content is increased up to 14% (Fig. 8). Both steels recrystallized to a {225} ⟨533⟩ texture (Fig. 9).

The increase in intensity of {110} ⟨112⟩ and {110} ⟨001⟩ rolling-texture components at 600°C (Fig. 10) has been caused by recovery, since the small amount of martensite which has transformed at this temperature would be insufficient to account for such an increase in texture intensity. The {225} ⟨533⟩ recrystallization component appears above 600°C. These results are amplified by the 220γ line profiles (Fig. 11), which show that recovery within the austenite starts at ~500°C, becomes much more pronounced at 600°C, but is followed by recrystallization between 600°C and 700°C.

Discussion

Formation of recrystallization textures from the duplex (α' + γ) rolling textures of metastable austenitic steels involves several stages. These are:

1. recovery within the deformed austenite,
2. recovery within the deformed martensite,
3. formation of recrystallization nuclei within the range of austenite rolling-texture orientations,
4. martensite to austenite shear transformation,
5. variant selectivity during the transformation,
6. recovery within the derived austenite,
7. formation of recrystallization nuclei within the range of transformed orientations,
8. competitive growth of favourable orientations to form the final texture: this is influenced by the relative amounts of α' and γ formed during rolling.
Successive stages overlap, but the role of each can be discussed separately, in the general order of occurrence.

Recovery merely leads to sharpening of the existing texture about the principal orientations. This has been illustrated for both 18%Cr-10%Ni and 18%Cr-14%Ni extremes of the composition range of metastable alloys. Recovery within the martensite prior to transformation has an important bearing on the derived austenite texture. The effect is that, even with random selection of variants during the transformation, that part of the austenite texture derived from the martensite should be less diffuse.

The martensite transformation in stainless steels has been the subject of many research papers, and there has been much discussion on the reactions which seem possible, i.e. whether \( \gamma \rightarrow \alpha' \) or \( \gamma \rightarrow \varepsilon \rightarrow \alpha' \). It is generally considered that \( \varepsilon \) is a transition phase and that continuing deformation will transform it to \( \alpha' \) martensite, so that the final product is the same in either case. The reverse transformation is simply \( \alpha' \rightarrow \gamma \). In relation to development of polycrystalline textures it is the orientation relationships between \( \gamma \) and \( \alpha' \) phases that are of interest rather than the mechanism of the transformation itself. Both the Kurdjumov–Sachs and Nishiyama relationships have been reported,

in the former \( (111)_\gamma || (110)_{\alpha'} \) and \( [1\bar{1}0]_\gamma || [1\bar{1}1]_{\alpha'} \)
in the latter \( (111)_\gamma || (110)_{\alpha'} \) and \( [\bar{2}11]_\gamma || [1\bar{1}0]_{\alpha'} \).

The two orientations are related by a \( 5^\circ 16' \) rotation of the b.c.c. lattice about the \( [1\bar{1}0]_\gamma \) axis. This difference between the two relationships is sufficiently small to allow them to yield similar distributions of variant orientations after transformation and in view of the fact that the consensus of opinion favours the Kurdjumov–Sachs relationship, the present discussion will be confined to interpretation of results in terms of this mechanism.

To interpret the \( \gamma \) texture immediately after reverse transformation, the martensite reaction will first be considered purely from a geometrical point of view, i.e. with complete random selection of variants. The results of the analysis are presented in Table 3 and Fig. 12.

<table>
<thead>
<tr>
<th>Principal components of ( \alpha' ) rolling texture</th>
<th>Components resulting from ( \alpha' \rightarrow \gamma )</th>
<th>Intensities relative to original</th>
</tr>
</thead>
<tbody>
<tr>
<td>{001} \langle 1\bar{1}0\rangle</td>
<td>{100} \langle 001\rangle</td>
<td>( \frac{1}{2} )</td>
</tr>
<tr>
<td>{111} \langle 1\bar{1}2\rangle</td>
<td>{110} \langle 5\bar{5}7\rangle</td>
<td>( \frac{1}{3} )</td>
</tr>
<tr>
<td>{112} \langle 1\bar{1}0\rangle</td>
<td>{4\cdot3} \langle 3\cdot4\rangle</td>
<td>( \frac{1}{2} )</td>
</tr>
<tr>
<td>{4\cdot3} \langle 6\cdot8\cdot1\cdot7\rangle</td>
<td>{6\cdot8\cdot1\cdot7} \langle 3\cdot7\cdot5\rangle</td>
<td>( \frac{1}{3} )</td>
</tr>
</tbody>
</table>

The \{001\} \langle 1\bar{1}0\rangle_{\alpha'} \) component thus gives rise to one group of orientations centred on \{100\} \langle 001\rangle, and two groups centred on twin-related \{110\} \langle 5\bar{5}7\rangle_{\gamma} \) orientations [Fig. 12(a)].

The \{111\} \langle 1\bar{1}2\rangle_{\alpha'} \) texture consists of two twin-related components, each of which gives rise to one group of orientations centred on \{4\cdot3\} \langle 3\cdot4\rangle, and two groups centred on twin-related \{4\cdot3\} \langle 6\cdot8\cdot1\cdot7\rangle_{\gamma} \) components. The combined effect [Fig. 12(b)] produces two twin-related \{4\cdot3\} \langle 3\cdot4\rangle \) components and four components of the type \{4\cdot3\} \langle 6\cdot8\cdot1\cdot7\rangle_{\gamma} \).

Similarly, the \{112\} \langle 1\bar{1}0\rangle_{\alpha'} \) texture consists of two twin-related components, each of which gives rise to
one group of orientations centred on \{430\} \langle001\rangle, and two groups centred on \{6,8,17\} \langle575\rangle, with the combined effect [Fig. 12(c)] of two \{430\} \langle001\rangle components and four components of the type \{6,8,17\} \langle575\rangle.

Fig. 12(d) is a prediction of the resultant austenite texture in terms of these ideal orientations. Obviously this Figure is not meant to be a full description of the texture, but if the analysis is valid these orientations should at least fall within the general spread. Comparison with Fig. 5(a) shows that the \(\gamma\) texture of 18\% Cr-10\% Ni steel after \(\frac{1}{2}\) hour at 600\(^\circ\)C generally conforms to the predictions, but the group of orientations centred on \{100\} \langle001\rangle, and originating from \{001\} \langleT10\rangle, is missing. This is direct evidence of preferential selection of variants; its effect is to produce a sharper transformation texture than would arise when random selection occurs. Suppression of certain habit planes during transformation has previously been reported by Bokros & Parker (1963) and Borik & Richman (1967).

Subsequent recrystallization in the austenite depends

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**Fig. 5.** (200)\(\gamma\) pole figures of 18\% Cr-10\% Ni steel cold rolled to 95\% reduction and annealed; (a) \(\frac{1}{2}\) hour at 600\(^\circ\)C, air cooled, (b) \(\frac{1}{2}\) hour at 700\(^\circ\)C, air cooled, (c) \(\frac{1}{2}\) hour at 900\(^\circ\)C, air cooled.
on the amounts of $\alpha'$ and $\gamma$ in the cold-rolling texture. In the more stable alloys, nuclei forming within the range of $\gamma$ rolling-texture orientations are required to grow through a fairly limited range of matrix components in order to become part of the recrystallization texture. In highly metastable alloys, however, the wider range of orientations of the b.c.c. rolling texture is transferred via the $\alpha' \rightarrow \gamma$ transformation to the derived austenite texture. From Fig. 5(a), there are two main distributions involved, but there is also a considerable spread of orientations between these high intensity areas. In order to become part of the recrystallization texture, nuclei must grow through a much wider range of matrix components than in the former case. The process is still further complicated when some austenite remains in the rolling texture since nuclei from this range of orientations will compete with those from the derived austenite. Density of nuclei, being governed by the relative amounts of the various orientation distributions, is highly influential in governing the final form of the recrystallization texture.

Following the principles set out by Dillamore (1964), it will be assumed that, in general, a nucleus will be favoured for growth if it has a (111) pole almost in common with the matrix into which it grows and if the rotation about this pole is $\sim 40^\circ$. Tables 4 to 7 compare geometrical relationships of the principal recrystallization components for different ranges of matrix components in terms of $\theta$, the angle of misfit between (111) poles of nucleus and matrix, and $\phi$, the rotation about this almost common (111) pole. Comments regarding growth potential are generalized and consider not only the geometrical relationships but also the number of matrix components through which each nucleus must grow.

### Table 4. Geometrical conditions governing growth of \{225\}(533) nuclei

<table>
<thead>
<tr>
<th>Parent {110} (T12)</th>
<th>$\theta$</th>
<th>$\phi$</th>
<th>Comments</th>
</tr>
</thead>
<tbody>
<tr>
<td>{110} (001)</td>
<td>10</td>
<td>26</td>
<td>Favourable</td>
</tr>
<tr>
<td>{110} (557)</td>
<td>8</td>
<td>26</td>
<td>Favourable</td>
</tr>
<tr>
<td>{430} (001)</td>
<td>9</td>
<td>41</td>
<td>Favourable</td>
</tr>
<tr>
<td>{430} (6,8,17)</td>
<td>10</td>
<td>37</td>
<td>Partially favourable</td>
</tr>
<tr>
<td>{430} (340)</td>
<td>23</td>
<td>37</td>
<td>Unfavourable</td>
</tr>
<tr>
<td>{6,8,17} (375)</td>
<td>3</td>
<td></td>
<td>Close to {225} (533)</td>
</tr>
</tbody>
</table>

For a f.c.c. deformation texture of the \{110\} (T12) type, Dillamore (1964) deduced that nuclei favourable for growth from grain boundaries or deformation-band boundaries would be of the types \{4,9,17\} (385), \{490\} (001) and \{112\} (T10). The rolling texture also contains a minor \{110\} (001) component which, as shown by Fig. 10, occupies about one quarter of the total volume. Both \{112\} (T10) and \{490\} (001) nuclei are

![Fig. 6. Variation of intensities of \gamma texture components during annealing of 18% Cr-10% Ni steel.](image-url)
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Fig. 7. 220y line profiles for 18% Cr–10% Ni steel.

Table 6. Geometrical conditions governing growth of \{6,10,1\}\{531\} nuclei

<table>
<thead>
<tr>
<th>Parent</th>
<th>(\theta)</th>
<th>(\varphi)</th>
<th>Comments</th>
</tr>
</thead>
<tbody>
<tr>
<td>{110}{\langle112\rangle}</td>
<td>12(^\circ)</td>
<td>40(^\circ)</td>
<td>Favourable</td>
</tr>
<tr>
<td>{110}{001}</td>
<td>12(^\circ)</td>
<td>38(^\circ)</td>
<td>Unfavourable</td>
</tr>
<tr>
<td>{110}{\langle557\rangle}</td>
<td>18(^\circ)</td>
<td>60(^\circ)</td>
<td>Unfavourable</td>
</tr>
<tr>
<td>{430}{001}</td>
<td>5(^\circ)</td>
<td>33(^\circ)</td>
<td>Partially favourable</td>
</tr>
<tr>
<td>{430}{\langle6,8,17\rangle}</td>
<td>20(^\circ)</td>
<td>40(^\circ)</td>
<td>Partially favourable</td>
</tr>
<tr>
<td>{430}{\langle340\rangle}</td>
<td>13(^\circ)</td>
<td>43(^\circ)</td>
<td>Partially favourable</td>
</tr>
<tr>
<td>{430}{\langle6,8,17\rangle}</td>
<td>10(^\circ)</td>
<td>47(^\circ)</td>
<td>Partially favourable</td>
</tr>
<tr>
<td>{430}{\langle340\rangle}</td>
<td>14(^\circ)</td>
<td>25(^\circ)</td>
<td>Favourable</td>
</tr>
<tr>
<td>{6,8,17}{\langle375\rangle}</td>
<td>12(^\circ)</td>
<td>45(^\circ)</td>
<td>Unfavourable</td>
</tr>
</tbody>
</table>

When a small amount of \(\alpha\)' martensite is formed during cold rolling, as in the 18%Cr–14%Ni steel, \{225\}\{\langle333\rangle\} component is confronted with other ranges of matrix orientations, which are also represented in Table 4. The \{6,8,17\}\{\langle375\rangle\} component is very close to \{225\}\{\langle333\rangle\} and therefore merely becomes part of the range of growing nuclei. Small amounts of this component will actually enhance development of the recrystallization texture by spontaneously providing suitable nuclei.
texture can no longer be considered entirely random. Orientations such as \(\{110\} \langle 557\rangle\) and \(\{430\} \langle 6,8,17\rangle\) are close to \(\{110\} \langle T12\rangle\), and may be considered as part of the spread around \(\{110\} \langle T12\rangle\). Similarly \(\{430\} \langle 001\rangle\) may be considered as part of the spread around \(\{110\} \langle 001\rangle\). Matrix components such as these present no serious inhibiting effects, as shown by their geometrical relationships (Table 4); at least one out of two or two out of four matrix components remain favourable for \(\{225\} \langle 553\rangle\) nuclei. However, the greater the deviation from the \(\{110\} \langle T12\rangle - \{110\} \langle 001\rangle\) spread, then the greater is the likelihood of inhibition occurring. Orientations of the \(\{430\} \langle 340\rangle\) type will present the greatest inhibiting effect, due to a very large angle of misfit of both components with \(\{225\} \langle 553\rangle\) (Table 4). For small amounts of \(\alpha'\), the effect will be negligible, but increasing metastability will eventually restrict development of the \(\{225\} \langle 553\rangle\) recrystallization texture through the presence of \(\{430\} \langle 340\rangle\) type matrix components and others. The present results show that at least 30% \(\alpha'\) can be tolerated before this occurs.

For the opposite extreme of austenite stability, i.e. 18%Cr–10%Ni steel, the recrystallization texture consists mainly of two groups of orientations, centred on \(\{230\} \langle 3,2,14\rangle\) and \(\{6,10,1\} \langle 531\rangle\), with spread towards \(\{6,8,17\} \langle 575\rangle\) type minor components. Nuclei of these orientations can originate entirely from the range of orientations produced by the \(\alpha' \rightarrow \gamma\) shear transformation. Table 5 shows that conditions are highly favourable for \(\{230\} \langle 3,2,14\rangle\) nuclei to grow into the \(\{110\} \langle 001\rangle + \{430\} \langle 001\rangle\) matrix distribution. Other orientations have at least one component which tends to inhibit growth by virtue of its mismatch with the matrix. Taking an overall view, \(\{230\} \langle 3,2,14\rangle\) nuclei seem to be capable of absorbing or partially absorbing all orientations except those of the type \(\{6,8,17\} \langle 575\rangle\). Table 6 shows that nuclei of the type \(\{6,10,1\} \langle 531\rangle\) would tend to be inhibited by the \(\{110\} \langle 001\rangle + \{430\} \langle 001\rangle\) spread. Apart from this, these nuclei also seem capable of absorbing or partially absorbing the matrix, with the exception of \(\{6,8,17\} \langle 575\rangle\). Particular differences (Table 6) are that the two \(\{110\} \langle T12\rangle\) components are now equally favourable, and growth into both of the \(\{430\} \langle 340\rangle\) components is likely, although the angle of rotation is less favourable for one of them. The \(\{6,8,17\} \langle 575\rangle\) orientation is much less favourably oriented for the range of matrix components (Table 7). In particular, \(\{430\} \langle 340\rangle\) type matrix components and others. The present results show that at least 30% \(\alpha'\) can be tolerated before this occurs.

![Fig. 9. (200)y pole figure of 18%Cr–14%Ni steel cold rolled and annealed (95% reduction; \(\frac{1}{2}\) hour at 900°C, air cooled).](image)

![Fig. 10. Variation of %\(\alpha'\) and intensities of \(\gamma\) texture components during annealing of 18%Cr–14%Ni steel.](image)

![Fig. 11. 220\(\gamma\) line profiles for 18%Cr–14%Ni steel.](image)
components will tend to inhibit growth. The \(\{6,8,17\}\) orientation is not easily absorbed by \(\{230\}\) orientation (Table 5) or by \(\{6,10,1\}\) orientation (Table 6) so that the spread towards this orientation in the final texture [Fig. 5(c)] probably results from its retention by ‘recrystallization in situ’.

Nuclei arising from the small amount of austenite in 18\%Cr–10\%Ni steel will merely become part of the spread centred on \(\{6,8,17\}\) orientation and therefore have no significant effect on the recrystallization texture. Increasing amounts of austenite will increase the density of \(\{225\}\) nuclei until they eventually provide major competition with other growing nuclei. Transition to this type of recrystallization texture will take place as the austenite stability increases.

Formation of the \(\{230\}\) + \(\{6,10,1\}\) recrystallization texture cannot be interpreted unambiguously. In general, all that can be concluded is that recrystallization takes place by growth competition between \(\{230\}\) and \(\{6,10,1\}\). The fact that there are two main types of nuclei involved will in one sense help the growth process, i.e. grains unfavourably

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Fig. 12. Positions of \(\{100\}\) poles after transformation from principal components of the \(\alpha'\) rolling texture. (a) \(\{001\}\) \(\langle\mathbf{T}10\rangle\alpha' \rightarrow \gamma\), (b) \(\{111\}\) \(\langle\mathbf{T}12\rangle\alpha' \rightarrow \gamma\), (c) \(\{112\}\) \(\langle\mathbf{T}10\rangle\alpha' \rightarrow \gamma\), (d) \(\gamma\) texture in terms of ideal orientations.
oriented for one nucleus might be favourably oriented for the other. From the present analysis, it is likely that the \{110\} \(\langle 001 \rangle\) + \{430\} \(\langle 001 \rangle\) spread disappears by growth of the \{230\} \(\langle 3,2,14 \rangle\) component. The low angle of misfit of the latter with the matrix will actually favour rapid growth at an early stage during recrystallization. This probably accounts for a large part of the texture rearrangement which takes place between 600°C [Fig. 5(a)] and 700°C [Fig. 5(b)] when the intensity of \{110\} \(\langle 001 \rangle\) has rapidly decreased. The \{6,10,1\} \(\langle 531 \rangle\) nuclei are more favourably oriented for growth into \{110\} \(\langle 112 \rangle\) and \{430\} \(\langle 340 \rangle\) components, and probably account for the greater proportion of growth into this part of the matrix.

Relative contributions of the various stages may be summarized as follows:

Recovery is important in that it leads to sharpening of both \(\alpha'\) and \(\gamma\) rolling textures. When large amounts of martensite are formed during rolling, its effect is more significant, i.e. texture sharpening produced in the martensite prior to reverse transformation will contribute to a less diffuse spread of orientations in the derived austenite than if recovery was delayed until after the transformation. The crystallographic nature of the \(\alpha' \rightarrow \gamma\) transformation limits the spread of orientations which are present in the austenite immediately after transformation, and variant selectivity can contribute to texture sharpening at this stage by suppression of certain variants of the Kurdjumov-Sachs orientation relationship. As the alloy stability increases and less martensite is formed, the influence of the transformation becomes less important. Throughout the range of metastable austenites, oriented growth rather than oriented nucleation is largely responsible for development of recrystallization textures. However, the \(\alpha' \rightarrow \gamma\) transformation is capable of spontaneously producing suitably oriented nuclei and certainly in the 18%Cr-10%Ni alloy the mechanism should be partly considered as oriented nucleation.

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References


Glassy Carbon as Standard for the Normalization of Small-Angle Scattering Intensities

BY R. PERRET* AND W. RULAND†

Union Carbide European Research Associates, S.A., Rue Gatti de Gandon 95, 1180 Bruxelles, Belgium

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The basic conditions for the normalization of X-ray small-angle scattering for pin-hole collimation, infinite and finite slit heights are reviewed. The characteristics of glassy carbon are discussed in the light of these conditions. It is shown that glassy carbon has a number of advantages which makes it an excellent choice as a standard sample.

In the course of our recent studies on the small-angle scattering of glassy carbon (Perret & Ruland, 1971) it occurred to us that this material would be excellent as standard samples for the normalization of small-angle scattering intensities. The purpose of this note is to give a short description of the characteristics of this material and its advantages over other types of standard samples.

Before discussing the characteristics of the material we should like to define what we consider the essential relationships to be applied in the normalization procedure for the scattering of a sample of arbitrary structure.

Pine-hole collimation

Assuming a sample with a linear absorption coefficient \(\mu\) in the form of a platelet of thickness \(t\) the scattering intensity \(I_{obs}\) is given by

\[ I_{obs}(s_{13}) = vt e^{-\mu t} I_0(s_{13}) \cdot I_{e.u.}(s_{13}), \]

\(v = 7.947 \cdot 10^{-10} \text{[A}^4]\).

\(s_{13}\) is the component of the vector \(s(s = 2 \sin \theta /\lambda)\) in the tangent plane to the Ewald sphere at the origin of the reciprocal space. \(I_0(s_{13})\) is the image of the primary beam in this plane as measured under the conditions

* Present address: Centre de Recherches de Kléber-Colombes, 49 rue Jean Jaurès, 95-Bezons, France.
† Present address: 61 avenue Kersbeck, 1190-Bruxelles, Belgium.