1. Introduction

During the conventional hot strip rolling process, C–Mn steels are subjected to finishing deformations in the single phase austenitic region. Currently, due to a number of technological benefits, there is growing interest in the application of finishing deformations after the onset of the austenite to ferrite transformation. However, whilst the processes of deformation, restoration and transformation in and from the single phase austenite are relatively well understood, deformation of austenite/ferrite mixtures and the subsequent transformation of the remaining austenite fraction have received very little attention. In particular, the dominant mechanisms controlling the evolution of complex and widely different microstructures have yet to be clearly defined.

With respect to the latter, complex strain path deformation may prove to be an invaluable tool. It has recently been reported by several authors that on the application of a fully reversed deformation (for instance an anti-clockwise torsional deformation following a clockwise torsional deformation or a tensile deformation following a compressive deformation) that the grain aspect ratio introduced by the forward deformation is anulled during the reversal, leading to a restoration of the original polygonal grain shape. However, whilst the processes of deformation, restoration and transformation in and from the single phase austenite are relatively well understood, deformation of austenite/ferrite mixtures and the subsequent transformation of the remaining austenite fraction have received very little attention. In particular, the dominant mechanisms controlling the evolution of complex and widely different microstructures have yet to be clearly defined.

In this study the effect of complex strain histories on the deformation response and the evolution of microstructure in intercritically deformed C–Mn steel has been considered. A number of experiments are described in which mixed austenite/ferrite structures have been subjected to both unidirectional and reversed torsion. It is shown that such an approach, in combination with rigorous characterisation of the final microstructure, enables the key mechanisms responsible for the evolution of microstructure to be more clearly identified. Perhaps most significantly, manipulation of the strain history has been shown to be a method that enables the separation of grain geometry and deformation substructural effects on softening and transformation mechanisms. Due to its wider methodology adopted for the characterisation of the microstructures of intercritically deformed steels and implicit determination of the set of metallurgical processes resulting in the observed microstructures is also described in detail.

2. Experimental Methods

2.1. Materials

The chemical analysis of the C–Mn steel chosen for this investigation is given in Table 1. This steel was chosen principally because it has been the subject of previous extensive investigations which have lead to a detailed understanding of the deformation and transformation in and from single phase austenite.

| Element | C  | Mn  | Si  | Al  | Cr  | Ni  |
|---------|----|-----|-----|-----|-----|-----|
| Weight %| 0.08 | 0.51 | 0.51 | 0.41 | 0.17 | 0.12 |
2.2. Mechanical Testing

All experiments were carried out on a Bähr 810 combined axial torsion deformation machine using samples of the geometry shown in Fig. 1. This machine allows for tensile, compressive and shear deformation at temperatures up to 1250°C. In these tests only the shear deformation (i.e. torsion) mode was utilised. The torsional deformation was imposed using a hydraulic system with a clutch/valve combination that makes possible very short reversal times (typically 20 ms). The angle of rotation is measured by a potentiometer on the torque drive shaft. The temperature is measured and controlled via a calibrated pyrometer. Heating is provided by an induction coil that has a second coil wound within it to provide direct He-gas quenching on the gauge section of the sample. In order to ensure that the temperature gradient from middle to end of the deforming length is of the order of 3°C, the coil itself extends beyond the gauge section and heats the shoulders as well. Balanced cooling of the clamps ensures that both ends of gauge section are at the same temperature. Visual inspection of the sample indicated that the deformation was distributed homogeneously over the gauge section, as was later confirmed by metallographic observation.

It is well known that the length of the gauge section can change during torsion. In cases where this happens the axial (i.e. non-twisting) clamp can either be held in a fixed position or allowed to float freely. In the current experiments the system continuously measured the force on the axial ram and maintained zero force, thus simulating a floating clamp. At the same time length changes during the deformation were continually measured. All length changes were attributed to the gauge section, assuming a constant cross-sectional area along the length of the gauge section, the system recalculated the gauge section radius and thus adjusted the rotational speed of the torque shaft to maintain a constant true strain rate.

When calculating shear stress from the torque the strain and strain rate sensitivity parameters of the material need to be known. Thus, in principle, multiple tests at different strain rates are necessary to derive the strain rate sensitivity parameter before any single flow stress curve can be calculated. However, the analysis of Barraclough9 showed that, for solid specimens, the solution for the stress is independent of the strain and strain rate sensitivity parameters at a radial position of 0.72r (where r is the outer radius of the gauge section). Solving at this radial position, the effective radius, means that the data from each test is self-sufficient, i.e. data from a matrix of tests is not needed to process the individual dataset. The dimensions of the gauge section at the start of the deformation cannot be known directly from the measured data, instead data from dilatometry samples of the same steel exposed to the same heating/cooling cycle are used to obtain an effective thermal expansion coefficient. This coefficient, in conjunction with the cold starting dimensions of the gauge section, is used in the post-test data analysis.

In all cases samples were austenitised at 900°C prior to cooling to the deformation temperature in the intercritical region at a rate of 4.5°C/s. In the series of experiments described here a single deformation temperature of 800°C, which equates approximately, to a structure comprising 70% ferrite and 30% austenite, has been used. The deformation schedule applied is illustrated schematically in Fig. 2.

The complete matrix of experiments conducted is presented in Table 2. From this table it can be seen that three basic forms of deformation history have been imposed. In the simplest case a single forward deformation (+) has been applied. Two-step deformations involving two sequential forward deformations have also been performed (+/-). In some cases the deformation steps were separated by a 100 s delay. In the case of the remaining tests a two step forward/reverse deformation schedule (+/-) has been imposed. All samples were quenched with He gas after the termination of the deformation schedule. The average cooling rate over the 800–500°C interval was approximately 13.5°C/s.

3. Metallography

Subsequent to testing all samples were prepared for metallographic examination. All metallography was carried out at the effective radius (i.e. 0.72r). A constant radial location was chosen in order to ensure that direct comparison of the
structures observed in all samples could be made whilst avoiding errors which may arise due to the radial variation in imposed strain and strain rate which result from deformation in torsion. Sectioned samples were ground and polished in the usual manner and were then chemically etched in Marshals reagent. Etched samples were then examined using an Olympus BHX optical microscope using both dry and oil immersion objective lenses.

Image post processing and quantitative metallography was conducted using a Leica, Q-win® system. Prior to measurement, digitised images were post-processed to enhance grain boundary contrast. In some cases this task was performed electronically but in cases where the structure was too complex or where the boundary etch was not uniform (as was typically observed for heavily deformed structures comprising deformation substructures) it proved necessary to trace the structure on to an acetate sheet by hand and to scan the trace to create a digitised grey-scale image. The grain boundaries were then detected and a binary image of the grain boundary structure constructed. From such binary images detailed grain size distribution data could be obtained.

4. Results

4.1. Flow Stress Data

The flow stress data from the torsion and reversed torsion tests are shown in Fig. 3(a): The single forward deformation to $\varepsilon = 1.2$ is the baseline against which all others should be compared. The results show that in an interrupted forward-forward test ($\varepsilon = +0.6$ and $+0.6$) the stress returns immediately to the steady state stress level, indicating recovery did not occur between the two deformations. The forward-reverse test ($\varepsilon = +0.6$ and $-0.6$, with a reversal time of $\sim 25$ ms) indicates a short transient in the work hardening, but with the flow stress returning to the same steady state level as the baseline curve. The fact that the steady state stress is the same for the reversed and simple forward cases would suggest some parity of microstructures in the two cases, however this need not be so.

The data analysis showed that the strain rate control during the different tests is very good and, apart from transients at start-up, remains at a constant $0.75$ s$^{-1}$ (Fig. 3(b)). The flow stress curves from the various tests indicate a high level of repeatability (Fig. 3(a)). This confirms that the thermo-mechanical history of the metallographic samples was as programmed and adds a level of confidence to the analysis.

4.2. Metallographic Observations

Illustrations of the microstructures resulting from the deformation schedules imposed are shown in Figs. 4–9. Micrographs showing the microstructure typical the re-heated and undeformed samples are shown in Fig. 4. It is clear from these figures that the resulting structure in this case is one of uniformly sized, polygonal ferrite.

In the case of samples subjected to a forward deformation only, as can be seen from Figs. 5 and 6, the resulting microstructure comprises deformed ferrite grains with an aspect ratio and well-defined internal substructure together with polygonal ferrite. It should be stressed that, due to the fact that Marshals reagent reveals high and low angle grain boundaries alike, distinguishing between subgrains and fine transformed polygonal ferrite grains is made difficult. In the qualitative sense, subgrains may only be positively identified in cases when they clearly belong to a colony enclosed within an elongated parent grain. The alignment of the pearlite colonies provides a clear illustration of the global deformation since they become inclined to the sample axis with an angle which is dependent upon the applied strain. Comparison of the micrographs shown in Figs. 5 and 6 indicates that, apart from a clearly differing inclination of the pearlite colonies, samples deformed to final strains of both 0.6 and 1.2 yield structures which are only subtly different. However, it would appear that the deformed ferrite contains a more clearly defined substructure in the case of the more heavily deformed material (Fig. 6). This leads to even greater difficulty in distinguishing between well-defined subgrains and fine transformed ferrite grains. Interrupted forward-forward deformation with insignificant delay between the deformation steps yields a microstructure indistinguishable from that resulting from a single forward deformation to the same aim strain. Such structures are illustrated in Fig. 7.

The microstructures of materials subjected to a fully reversed deformation are illustrated in Figs. 8 and 9. Comparison of these micrographs with those shown in Fig. 5 clearly illustrates that a restoration of the original,
equiaxed, polygonal grain morphology occurs on the imposition of a complete strain reversal. The resulting structure comprises of entirely polygonal ferrite. However, in this case a fraction of the ferrite grains can be seen to contain a substructure. As for the forward deformations, the principal effect of increasing the imposed strain lies in the scale of the structure and the definition of the substructure (compare Figs. 8 and 9). It would appear that the internal structure in the deformed ferrite fraction is more clearly defined after
the imposition of the higher strain and that the undeformed polygonal ferrite fraction is finer. Once again it becomes difficult to distinguish subgrains from fine transformed grains with the absence of elongated ferrite grains posing yet further problems.

4.3. Microstructural Predictions

A superficial attempt to interpret the microstructures observed revealed that no definitive conclusions could be drawn regarding the processes that had occurred. Microstructures were found to comprise of a number of components including polygonal ferrite, high aspect ratio ferrite and regions of fine grains or subgrains of the order of 1–3 μm in diameter. Whilst the individual components could be readily identified the existence of each component could be equally well attributed to a number of possible mechanisms. For instance polygonal ferrite could result either from transformation of the austenite fraction or from recrystallisation of deformed ferrite.

In an attempt to clarify the situation a systematic methodology was adopted for the interpretation of these structures. This involved first tabulating all mechanisms likely to play a role and subsequently conducting a series of thought experiments to predict the microstructures which would form as the result of combinations of these mechanisms. The predicted microstructures were found to fall into five principal categories, which in turn comprised as many as four subcategories. The experimental microstructures where then characterised qualitatively in terms of their component parts and compared to the predicted variations. At this stage the experimental microstructures could be assigned to a principal category and a number of mechanistic combinations could be eliminated. The experimental microstructures where subsequently subjected to a more rigorous quantitative examination (grain size determination, volume fraction analysis etc.) in order to identify the more subtle distinctions between the sub categories and to more clearly identify the controlling mechanisms.

Table 3 lists the mechanisms considered and shows the combinations of mechanisms held possible for each of the experimental deformation schedules imposed (0 indicates that a mechanism is switched off whilst 1 indicates that a mechanism is in operation). In the case of the purely reheated sample all mechanisms associated with deformation or post deformation phenomena can be switched off and the resulting predicted microstructure is one of purely polygonal ferrite with a uniform size distribution (i.e. variant A). The situation becomes increasingly complex for the series of deformation schedules considered and progressively more mechanisms must be taken into account.

The first mechanisms listed relate to the evolution of the deformed microstructure. In all, two processes are considered, namely a change in grain geometry (i.e. elongation of the grains) and the development of some internal substructure. In the case of unidirectional deformation these processes will inevitably occur in phase regions which accommodate strain. However, in the case of intercritical deformation, it is not clear if both phases do indeed undergo deformation. Here, several scenarios have been considered. In the first instance it has been assumed that the strain accommodated in the carbon rich austenite phase is negligible and that only the softer ferrite phase deforms. In the absence of post deformation mechanisms this would lead to a microstructure containing deformed ferrite and a strain free ferrite from post-deformation transformed austenite (variant B). Alternatively, in the event that the extent of deformation in austenite cannot be ignored (the likelihood of which increases with increasing strain), changes in austenite grain geometry and/or the introduction of an internal substructure may lead to a refinement of the polygonal ferrite fraction (variant D). The inclusion of a sub-variant, D’, which is qualitatively identical to variant D reflects the fact that quantifiable, strain dependant variations in microstructure may arise. For example, at higher applied strains, geometrical changes may be expected to be more extensive and the severity of substructure may be higher.

In the case of a fully reversed deformation, when the net strain approximates zero, two further complications may arise. In such cases it has been reported that, on reversal, restoration of the original grain geometry may occur and that partial unravelling of dislocation substructure formed during the forward deformation is also possible. During this exercise, two scenarios have again been considered. In the event that both processes occur fully the resulting microstructure would show no evidence of the prior deformation and consequently a uniform polygonal ferrite grain structure, corresponding to variant A, is predicted. In the event that reversal leads to the restoration of grain shape but not to the eradication of deformation substructure an additional variant, variant G, comprising transformed polygonal ferrite grains and polygonal ferrite grains containing a deformation substructure, would result. Again, separate situations are considered in which deformation is accommodated solely in the ferritic phase or in both the ferritic and austenitic phases. In the case that deformation is accommodated in both phases then transformation of the remaining austenite may give rise to a refinement of the transformed ferrite fraction leading to the inclusion of the sub-variant G’. That there may be a quantifiable variation in the scale of both the grain structure and the substructure with increasing total strain is reflected in the inclusion of two further sub-variants G'' and G*.

In addition to deformation mechanisms the restorative mechanisms of recovery and recrystallisation must also be accounted for. If sufficient deformation is accommodated within the ferrite fraction then the deformed ferrite may statically recrystallise. This could be envisaged to lead to a final structure comprising transformed polygonal ferrite and recrystallised polygonal ferrite. Qualitatively, this represents a structure identical to variant A (100% polygonal ferrite), however, a separate variant (variant C) must be included to account for the fact that the two processes may lead to polygonal ferrite fractions with quantifiably bimodal distribution of grain sizes. In the event that significant deformation is accommodated in the austenite phase static recrystallisation of the deformed austenite may conceivably occur prior to or during transformation. This gives rise to the microstructural variant E (and its higher strain sub-variant E’) which reflects the fact that this additional mechanism, which may result in further refinement of the final ferritic structure, could occur.

Finally, in a deformed austenite/ferrite structure under-
going transformation, ferrite nuclei which form on boundaries separating austenite and deformed ferrite grains may grow into both phases.\textsuperscript{3) That is to say that a driving force for the growth of the new phase exists not only in the austenite where a chemical driving force is present but also in the deformed ferrite where there is a stored energy of deformation. In effect the boundaries of the transforming regions have a mobility in both of the original microstructural components. This, so-called, transformation induced recrystallisation, may be expected to give rise to a fraction of abnormally coarse ferrite grains and a distinct microstructural variant, $F$.

It is clear from the above treatment that the resulting structures comprise several layers of complexity. This is illustrated in Tables 3 and 4. There are a total of six microstructural components possible, different combinations of which enable five qualitatively distinct microstructural categories to be distinguished. This from a total of seven microstructural variants which reflect the mechanistic combinations considered. Furthermore quantifiable differences may give rise to as many of four sub-variants in each case.

### 4.4. Comparison of Predicted and Real Microstructures

On the basis of the qualitative microstructural descriptions the resulting microstructures can be classified according to the scheme presented in the previous section. Table 5 provides a summary of the observations and lists the microstructural categories and variants which best fit the experimental observations. This qualitative comparison leads to the conclusion that the categories present are 1, 3 and 5 whilst the variants A, B, D and G occur. It would therefore appear, on the basis of these observations that a number of mechanisms can be ruled out. For example, there would appear to be no strong evidence that static recrystallisation of

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**Table 3.** Microstructural variations predicted as the result of the most likely mechanistic combinations.

| Strain History | $\alpha$ deformation | $\gamma$ deformation | Trans Induced SR | $\alpha$ SR | $\gamma$ SR | Resulting structure |
|----------------|-----------------------|----------------------|-----------------|------------|-----------|-----------------|
| 0              | 0 0 0 0 0 0 0         | 0 0                  | A               |            |           | Uniform PTrans  |
| 0.6            | 1 1 0 0 0 0 0         | 0 0                  | B               | Ar ss, PTrans |
|                | 1 1 1 0 0 0 1         | C                    | PTrans, $P_F$, possible bi-modal size distribution (50/50) |
|                | 1 1 1 1 0 0 0         | D                    | Ar ss, PTrans refined of A |
| 1.2            | 1 1 1 1 0 1 1         | E                    | PTrans refined of A, PRef refined of C |
|                | 1 1 1 1 1 0 1         | D'                   | Ar ss higher of B, PTrans refined of A |
|                | 1 1 1 1 1 0 1         | E'                   | PTrans refined of A, PRef refined of C |
|                | 1 1 1 1 1 1 0         | F                    | Ar ss, PTrans, Abnorm P |
| $\pm 0.3-0.3$  | 0 0 0 0 0 0 0         | A                    | Uniform PTrans |
|                | 0 1 0 0 0 0 0         | G                    | PTrans, P ss, Weekly ss ferrite |
|                | 0 1 0 1 0 0 0         | G'                   | PTrans slight refinement of G, PDef, Weekly ss ferrite |
| $\pm 0.6-0.6$  | 0 1 0 0 0 0 0         | G''                  | PTrans, $P_F$, stronger of G, Weekly ss ferrite |
|                | 0 1 0 0 0 1 0         | C                    | PTrans, $P_F$, possible bi-modal size distribution (50/50) |
|                | 0 1 0 1 0 0 0         | G'                   | PTrans refinement of G, P ss stronger of G, Weekly ss ferrite |
|                | 0 1 1 0 0 0 1         | C                    | PTrans, $P_F$, possible bi-modal size distribution (50/50) |

$P_F$ = Polygonal ferrite resulting from transformation from the austenite fraction.
$P_{Ref}$ = Polygonal ferrite resulting from recrystallisation of the deformed ferrite fraction.
$P_{ss}$ = Polygonal ferrite containing deformation substructure.
$A_r$ = Ascent ratio ferrite containing deformation substructure.
$Abnorm P$ = Abnormal or coarse Polygonal ferrite resulting from transformation induced recrystallisation.

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**Table 4.** The principal microstructural categories and their associated subcategories.

| Principal category | Description | Variants |
|--------------------|-------------|----------|
| 1                  | PTrans      | A        |
| 2                  | PTrans+PRef | E, F, C  |
| 3                  | PTrans+Ar ss| B, D, E  |
| 4                  | PTrans+Ar ss+Abnorm P | F |
| 5                  | PTrans+P ss | G, G', G'', G*** |

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**Table 5.** A summary of microstructural observations made on real deformed samples.

| Strain History | Observed microstructural components | Category | Possible Variants |
|----------------|-------------------------------------|----------|------------------|
| Reheat         | PT                                  | 1        | A                |
| $\pm 0.6$      | Arss, PFS                           | 3        | B, D, D'         |
| $\pm 0.6$      | Arss, P (coarse)                    | 3        | B, D, D'         |
| $+1.2$         | Arss, P (fine)                      | 3        | B, D, D'         |
| $\pm 0.6-0.6$  | P(100) fine, Arss                   | 3        | B, D, D'         |
| $\pm 0.3-0.3$  | P(100), P (big)                     | 5        | G, G', G'', G*** |
| $\pm 0.3-0.3$  | P(100), P (big), Arss               | 3        | B, D, D'         |
| $\pm 0.6-0.6$  | P(100), P (big)                     | 5        | G, G', G'', G*** |
| $\pm 0.6-0.6$  | P(100), P (big), Arss               | 3        | B, D, D'         |
| $\pm 0.6-0.6$  | P(100), P (big), Arss               | 5        | G, G', G'', G*** |
| $\pm 0.6-0.6$  | P(100), P (big)                     | 3        | B, D, D'         |
| $\pm 0.6-0.6$  | P(100), P (big), Arss               | 3        | B, D, D'         |
| $\pm 0.6-0.6$  | P(100), P (big), Arss               | 5        | G, G', G'', G*** |
| $\pm 0.6-0.6$  | P(100), P (big), Arss               | 3        | B, D, D'         |
the ferrite fraction occurs. This observation is strongly supported by the fact that imposition of a 100 s delay between deformation steps has no measurable effect on the resulting microstructure and that substructures observed clearly imply extensive recovery has occurred. These findings are in agreement with those of Wang\textsuperscript{11} who reported that dynamic recovery was the dominant restorative mechanism in a C–Mn steel deformed in the intercritical region to reductions of less than 85\%.

Since the substructure is not eradicated by a strain reversal, it would appear that unravelling of the substructure during reversal either does not occur or is not complete. This is in agreement with the observations of Valle et al.\textsuperscript{5} It also seems unlikely that transformation induced recrystallisation plays a significant role for those conditions imposed since abnormally coarse grains where not observed. However, without a more careful quantitative analysis it is not yet possible to distinguish whether deformation is accommodated in one or both phases or whether static recrystallisation of the austenitic fraction occurs prior to transformation. Such information may be obtainable from electron beam scattering diffraction (EBSD).

4.5. Quantitative Metallography

Grain size distributions for undeformed material and for material subjected a range of deformation schedules are shown in Figs. 10(a)–10(f). Since the technique used measures grain surface area, it is convenient to express grain size as an equivalent diameter. In this case the equivalent diameter represents the diameter of a circle with an identical surface area. It should also be noted that, as applied here, the technique can only yield data corresponding to the entire boundary network and that no distinction is made be-
tween high angle boundaries separating grains and low angle boundaries separating subgrains (Marshals reagent develops deformation substructure in ferrite\(^{10}\)). The grain size distributions therefore contain data corresponding to both grain and sub-grain populations and, in cases where the ferrite grain size may be expected to be fine, it is not possible to define a clear criterion for the separation of grains and sub-grains based on grain size. However, under most circumstances, one might expect transformed ferrite grains to be larger than subgrains and that the statistics at smaller grain sizes are dominated by the substructure. Similarly, the largest grain in the distribution could be assumed to be representative of the transformed ferrite fraction.

In the case of the re-heated and transformed material the grain size distribution is normal with a mean equivalent diameter of approximately 9.5 \(\mu\)m (see Table 4). Comparison of Figs. 10(a) and 10(b) shows that after the imposition of a simple forward deformation to a strain of 0.6 the distribution becomes skewed to smaller grain sizes (due to the introduction of a subgrain structure) and yields a refined mean equivalent diameter of approximately 3.7 \(\mu\)m (Table 4). The remaining Figs. (10(c)–10(f)) illustrate the effect of strain history and indicate that no significant change in grain size distribution occurs with either increasing applied strain or with the imposition of a strain reversal. Seemingly, the imposition of a delay between deformation steps also has no effect of grain size distribution (see Fig. 10(f)).

This observation would seem to suggest that the strain reversal does not give rise to a quantifiable change in the deformation substructure and is therefore in support of the earlier conclusion drawn from qualitative observations. Consideration of the mean grain sizes, plotted in Fig. 11(a), indicates that a further moderate refinement in mean grain size may be realised with increasing total applied strain. Comparison of Figs. 11(a) and 11(b) clearly reveals that mean grain size exhibits a dependency on total rather than net strain.

In an attempt to quantify the evolution of the transformed ferritic structure with strain history variations of the maximum, undeformed grain diameter have been measured. If it is assumed that, in the final structure, the ferritic grains containing no internal substructure can only result from transformation of the austenitic fraction (the only other possible source being recrystallisation of the ferritic fraction for which strong evidence to the contrary exists), then any dependency upon strain would indicate that significant strain accommodation in the austenite fraction occurs. The largest grain diameter was used in an attempt to avoid confusion between grains and subgrains. Figure 12 shows the variation of the diameter of the largest grain with total applied strain.

### Table 6. Mean grain sizes recorded for all deformation histories.

| Strain history | Equivalent grain diameter |
|----------------|--------------------------|
|                | Mean         | 95% Confidence limit |
| Re-heat        | 9.46         | ±0.26                |
| +0.6           | 3.67         | ±0.12                |
| +1.2           | 2.74         | ±0.11                |
| +0.6/+0.6      | 2.94         | ±0.11                |
| +0.3/+0.3      | 3.88         | ±0.29                |
| +0.6/+0.6      | 4.30         | ±0.52                |
| +0.6 100%      | 2.54         | ±0.16                |
| +0.6 100% +0.6 | 2.62         | ±0.14                |
| +0.6 100% −0.6 | 2.19         | ±0.26                |

![Fig. 11(a).](image1) The variation of mean equivalent grain diameter with total applied strain.

![Fig. 11(b).](image2) The variation of mean equivalent grain diameter with net applied strain.

![Fig. 12.](image3) Variation of the diameter of the largest, strain free, grain with total strain.
plied strain. The correlation between grain size and total strain would imply that strain is indeed accommodated in the austenite fraction.

5. Conclusions

Regarding the mechanisms of deformation: No evidence of recrystallisation of the ferrite fraction has been observed (heavily substructured ferrite was always present and a 100 s delay between deformations appears to have no effect). The deformation substructures observed in the ferrite fraction are indicative of rapid extensive recovery (well formed polygonal subgrain structures). Austenite accommodates significant strain (the grain size of the transformed fraction, here represented by the maximum grain size in the distribution, is strongly dependent on strain). Notwithstanding the difficulties associated with distinguishing grains from subgrains, consideration of the total grain size distribution implies that strain reversal does not lead to a change in the substructure (grain size distributions reveal a dependency on total not net strain), but it does lead to a restoration of grain shape. Evidence of transformation induced recrystallisation has not been observed (there is no evidence of abnormally large grains).

Regarding the metallographic technique: On its own, direct observation of the microstructure revealed little and raised more questions than it supplied answers. Quantification of the microstructures provided further insight into the possible microstructural events that took place during, and post, deformation. However, the technique of systematically positing all possible combinations of all possible events and deducing the subsequent hypothetical microstructures, on the basis of experience and extant literature, provided a great deal of otherwise unobtainable insight. By following this procedure and then cross-checking against independent characterisation of the real microstructures, many potential degrees of freedom were removed and greater certainty in the analysis was obtained.

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