On the Origin of Equiaxed Austenite Grains that Result from the Hot Rolling of Steel

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An investigation has been conducted into the influence of hot rolling on the microstructure of austenite in both a low alloy (transformable) and an austenitic stainless steel. Specimens of each steel were reheated and then given a single 50 pct reduction at various temperatures, after which the specimens were water quenched. The specimens were analyzed using optical, transmission electron and X-ray metallography. The results of these experiments showed that both steels exhibited equiaxed grains after high temperature rolling, elongated grains after low temperature rolling and a mixture of both types of grains after rolling at intermediate temperatures. One of the principal goals of this work was to study the origin of the equiaxed grains that result from high temperature rolling. Information obtained from dislocation observations and texture analysis has led to the conclusion that the mechanism most likely responsible for the equiaxed grains in as-rolled austenite is dynamic recrystallization.

AN observation frequently made during hot rolling experiments of low alloy steel is that when the deformation takes place at rather high temperatures, i.e. $\geq\sim$1000 °C (1832 °F), the prior-austenite grains that are observed upon quenching immediately after rolling exhibit an equiaxed morphology. It is also frequently observed that the equiaxed prior-austenite grains that are found after rolling are somewhat smaller than those present prior to rolling, i.e. the hot rolling had caused grain refinement of the austenite.[1-3] It is generally assumed that this new set of equiaxed prior-austenite grains has resulted from the recrystallization of the original grains, and that this recrystallization is a consequence of the hot rolling treatment. The principal goal of the work reported herein was to investigate the nature of the recrystallization which attends high temperature hot rolling and which is responsible for the observed grain refinement of the equiaxed prior-austenite.

There are four mechanisms that, individually, could cause an original set of equiaxed grains to be replaced by a second set during a thermomechanical process. These are: a) static recrystallization, b) dynamic recrystallization, c) metadynamic recrystallization, and d) continuous recrystallization. Static recrystallization occurs when both the nucleation and growth of the new grains take place after the deformation.[4] Dynamic recrystallization results when both the nucleation and growth take place during the deformation.[5,6] Metadynamic recrystallization obtains when the nucleation of the new grains occurs during deformation but the growth takes place after deformation.[7,8] Continuous recrystallization is somewhat different from the others in that it is not strictly a recrystallization process but rather an ad-

vanced form of recovery in which there is no gross movement of high angle grain boundaries.[9,10] Since there are several ways that hot rolling can lead to a final, equiaxed, prior-austenite grain morphology, it would appear quite tenuous to try to assign a mechanism to this process merely on the basis of observations of grain morphology made before and after rolling. The problem becomes even more complicated when considering the rolling behavior of low alloy steel austenite, since, in this case, observations must be made on the transformation products, ferrite or martensite, rather than on the phase of interest, the parent austenite.

In an attempt to address the question of which recrystallization mechanism is operative in association with hot rolling, it was decided to investigate the microstructural changes that attend high temperature rolling, and, in particular, to investigate the changes in defect structure and texture as well as grain size and shape. Since low alloy steel is not well suited for this purpose because the austenite, in this case, is stable only at high temperatures, a model material of AISI 304 austenitic stainless steel was considered for use. This material was considered because previous work had indicated that behavior of AISI 304 during hot rolling was very similar to that of low alloy steel austenite.[11]

To establish that AISI 304 could be used as a reasonable model for low alloy steel austenite in hot rolling studies, it was considered important to determine that the restoration processes that occurred during the hot deformation of the two types of steels were similar. This, of course, would mean that the metallurgical state of the austenite after hot rolling would be similar in the two types of steel. The metallurgical state of the AISI 304 austenite after hot rolling can be established through the application of standard optical, transmission electron microscopy (TEM) and X-ray techniques. However, since a low alloy steel is either ferrite or martensite when observed at room temperature, the metallurgical state of its prior-austenite can only be inferred from observations

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made on the transformation products. One of the most powerful methods of assessing the metallurgical condition of a thermomechanically treated steel specimen is to observe the texture exhibited by the specimen. For example, if the specimen is cold worked or recovered, its pole figure would indicate a deformation texture. If, however, the specimen had undergone recrystallization, the motion of the high angle boundaries that constitutes recrystallization would transform the deformation texture to a recrystallization texture. Hence, it is possible to learn a great deal about the metallurgical condition of a specimen by observing its pole figure.[12,13] Assessing the metallurgical condition of AISI 304 after hot rolling using pole figure analysis is straightforward since the as-rolled condition can be essentially retained by rapid quenching. However, assessing the metallurgical condition of the low alloy steel austenite after hot rolling through pole figure analysis is more complex because of the martensite transformation that occurs during the quench. The problem of assessing the metallurgical state of prior-austenite by pole figure analysis of the transformation products has been discussed by Jones and Walker[13] and others.[14] These studies have shown that it is possible to deduce the texture present in the austenite prior to transformation by observing the pole figure of the transformation products. The austenite and ferrite or martensite are known to be crystallographically related by orientation relationships typified by the Kurdjomov-Sachs orientation relationship: $\{111\}_{fcc}//\{110\}_{bcc}$; $<110>_{fcc}//<111>_{bcc}$.[15] As a result of this orientation relationship, the distribution of poles in the (111) pole figure of the parent austenite would be expected to be similar to the distribution of poles in the (110) pole figure of the ferrite or martensite. It is possible, therefore, to use the (110) ferrite or martensite pole figure to assess the metallurgical state of the parent austenite. This can be accomplished by using the texture of the ferrite or martensite to obtain the texture that had been present in the austenite prior to transformation.[13,14]

In summary, if it can be shown that AISI 304 steel and a low alloy steel exhibit similar behavior during hot rolling, e.g. similar optical microstructures and textures, then AISI 304 could be considered a valid model material for low alloy steel austenite and could be used in subsequent experiments to elucidate the recrystallization mechanism operative during high temperature hot rolling.

EXPERIMENTAL PROCEDURE

Two steels were employed in this study. One steel was AISI 304 and the second was a vanadium-aluminum-nitrogen-bearing microalloyed steel (V-steel). The nominal compositions of these steels are shown below.

	Composition, Wt Pct							
Steel	C	Mn	Si	Cr	Ni	Al	V	N
V-Steel	0.14	1.32	0.35	—	—	0.035	0.13	0.020
AISI 304	0.06	1.73	0.47	18.34	8.47	—	—	—

The AISI 304 steel was obtained as 12.7 mm (0.5 in.) thick, commercially produced hot rolled plate whereas the V-steel was produced, as an air melted 222.4N (50 pound) laboratory heat. After being homogenized at 1290 °C (2350 °F), the V-steel ingot was hot rolled to 16 mm (0.625 in.) plate and then air cooled to room temperature. The as-hot rolled condition represented the starting condition for each of the two steels for the subsequent hot rolling experiments.

Square specimens, 51 mm (2 in.) on a side, were machined from each steel and reheated to 1290 °C (2350 °F) for one hour. After reheating, the V-steel was given a 20 pct reduction at approximately 1200 °C (2200 °F) to embed a thermocouple at the center of the specimens and then the specimens were air cooled to the temperature of final hot rolling where they received a single 50 pct reduction. Because of excessive roll quenching, the processing of the 304 steel did not include the 20 pct reduction at 1200 °C (2200 °F). The 304 steel was reheated at 1290 °C and air cooled to the temperature of final rolling where it was given the single 50 pct reduction. The temperatures used for hot rolling varied between 1120 °C (2050 °F) and 790 °C (1450 °F). After hot rolling the specimens were either immediately water quenched or transferred to a furnace for isothermal holding prior to final water quenching.

Specimens for optical metallography were removed from the rolled material such that the longitudinal and short transverse directions could be observed. The specimens taken from the V-steel were mechanically polished using standard techniques and were then etched with a modified picral etch to reveal the prior-austentite boundaries. The specimens taken from the 304 steel were initially mechanically polished using standard techniques and were then electropolished in a solution that contained 630 ml methanol, 22.5 ml sulfuric acid and 3.5 ml distilled water. A voltage of approximately 28 V was employed with the electrolyte being maintained at −35 °C during polishing. The electropolished surface was then electroetched at room temperature in a solution of 10 gm oxalic acid and 100 ml water and at a voltage of about 3 V.

Specimens for TEM observation were prepared in two stages from suitably sectioned wafers that were removed from the rolled 304 steel. The first stage consisted of alternate chemical polishing and mechanical polishing. The chemical polishing employed a solution that contained 45 vol pct water, 30 vol pct nitric acid, 15 vol pct hydrochloric acid and 10 vol pct hydrofluoric acid. The second stage consisted of electropolishing in a jet polishing unit that used the same electrolyte that was described above for the preparation of specimens for optical metallography.

Specimens that were used in the determination of pole figures were removed from the mid-thickness of the rolled steels. These specimens were chemically polished to remove the effects of mechanical abrading. The surface parallel to the rolling plane was employed to determine the pole figures through the application of standard X-ray techniques.

RESULTS

The first step in the thermomechanical treatment of the two steels was a one hour hold at 1290 °C (~2350 °F). This reheating treatment produced equiaxed, recrystallized austenite in the two steels. The as-reheated grain diameter was approximately 200 microns (ASTM number 1–2) for the 304 steel and approximately 125 microns (ASTM number 2–3) for the V-steel. Other studies have shown that the austenite in each steel coarsens by way of a secondary recrystallization or abnormal grain growth mechanism.[11,16] The reheating treatment used placed each steel in the regime where the austenite grains had experienced complete secondary recrystallization.

The first goal of this program was to establish whether or not the 304 and the V-steels showed similar responses to simple hot rolling. The hot rolling behavior of the V-steel is shown in Fig. 1. When the V-steel was rolled at very high temperatures (≥1040 °C:~1900 °F), the resulting austenite exhibited an equiaxed morphology as shown in Fig. 1(a). When this steel was rolled at low temperatures (≤870 °C:~1600 °F), the resulting austenite had an elongated or "pancaked" morphology, Fig. 1(c). Rolling at intermediate temperatures led to a mixed austenite structure as shown in Fig. 1(b) for the V-steel rolled at 955 °C (~1750 °F).

The hot rolling behavior of the 304 steel is shown in Fig. 2. When the 304 steel was rolled at high temperatures (≥1075 °C:~2000 °F), the resulting structure exhibited a completely equiaxed morphology, Fig. 2(a). When the 304 steel was rolled at low temperatures (≤790 °C:~1450 °F), a microstructure consisting of elongated grains obtained, Fig. 2(c). When the rolling took place at temperatures between 1095 and 790 °C, a mixed grain morphology resulted as is shown in Fig. 2(b) for rolling at 1040 °C.

A comparison of the as-rolled microstructures shown in Figs. 1 and 2 reveals that the austenite in the two steels exhibited a similar variation in grain mor-
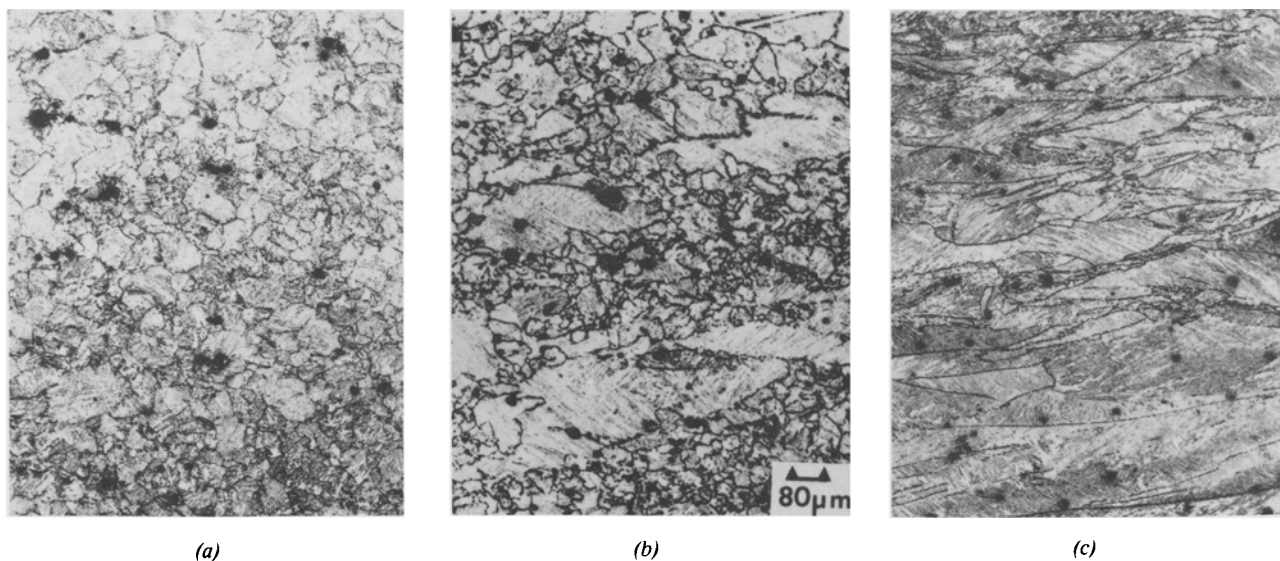

(a)　　　　　　　　(b)　　　　　　　　(c)

Fig. 1—Microstructure of prior-austenite of V-steel water quenched after one 50 pct rolling reduction at (a) 1040 °C, (b) 955 °C, (c) 870 °C.
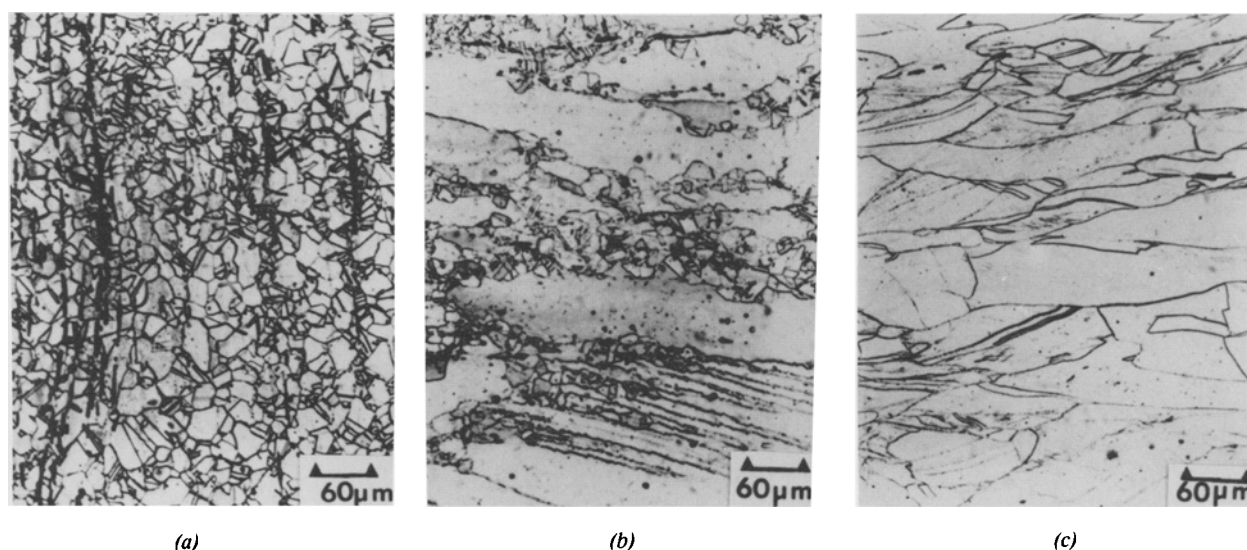

(a)　　　　　　　　(b)　　　　　　　　(c)

Fig. 2—Microstructure of AISI 304 steel water quenched after one 50 pct rolling reduction at (a) 1095 °C, (b) 1040 °C, (c) 790 °C.

phology with rolling temperature. That is, both steels have shown an equiaxed austenite morphology after high temperature rolling, an elongated grain morphology after low temperature rolling and a mixed grain morphology at intermediate rolling temperatures.

Pole figures were determined from specimens that showed either completely equiaxed as-rolled austenite grains, i.e. Fig. 1(a) for the V-steel and Fig. 2(a) for the 304 steel, or completely elongated grains, i.e. Fig. 1(c) for the V-steel and Fig. 2(c) for the 304 steel. The textures of the elongated austenites are shown in Figs. 3(a) and 4(a) for the V-steel and the 304 steel, respectively. In both cases the regions of highest pole density occurred in the general vicinity of orientations approximately 20 to 30 deg off the normal to the rolling plane and along the great circle connecting it with the rolling direction. The pole figures of the specimens with equiaxed austenite grains are presented in Figs. 3(b) and 4(b) for the V-steel and the 304 steel, respectively. These two pole figures show the regions of
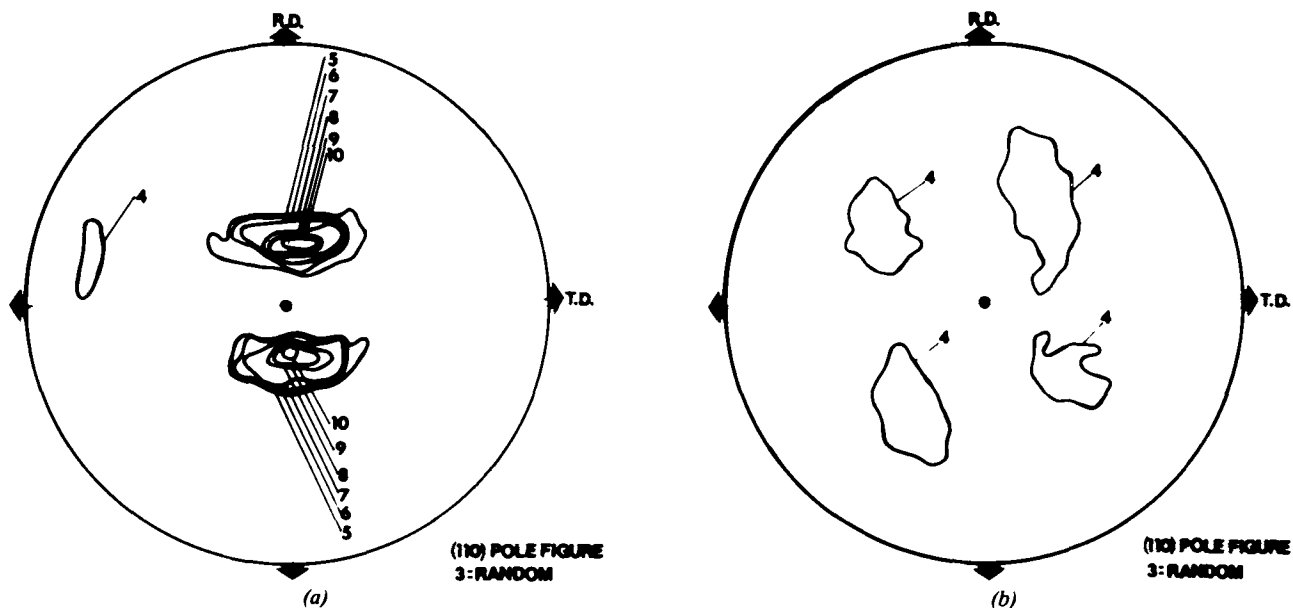

Fig. 3—(110) pole figures taken from V-steel (a) Elongated prior-austenite grains, (b) Equiaxed prior-austenite grains.
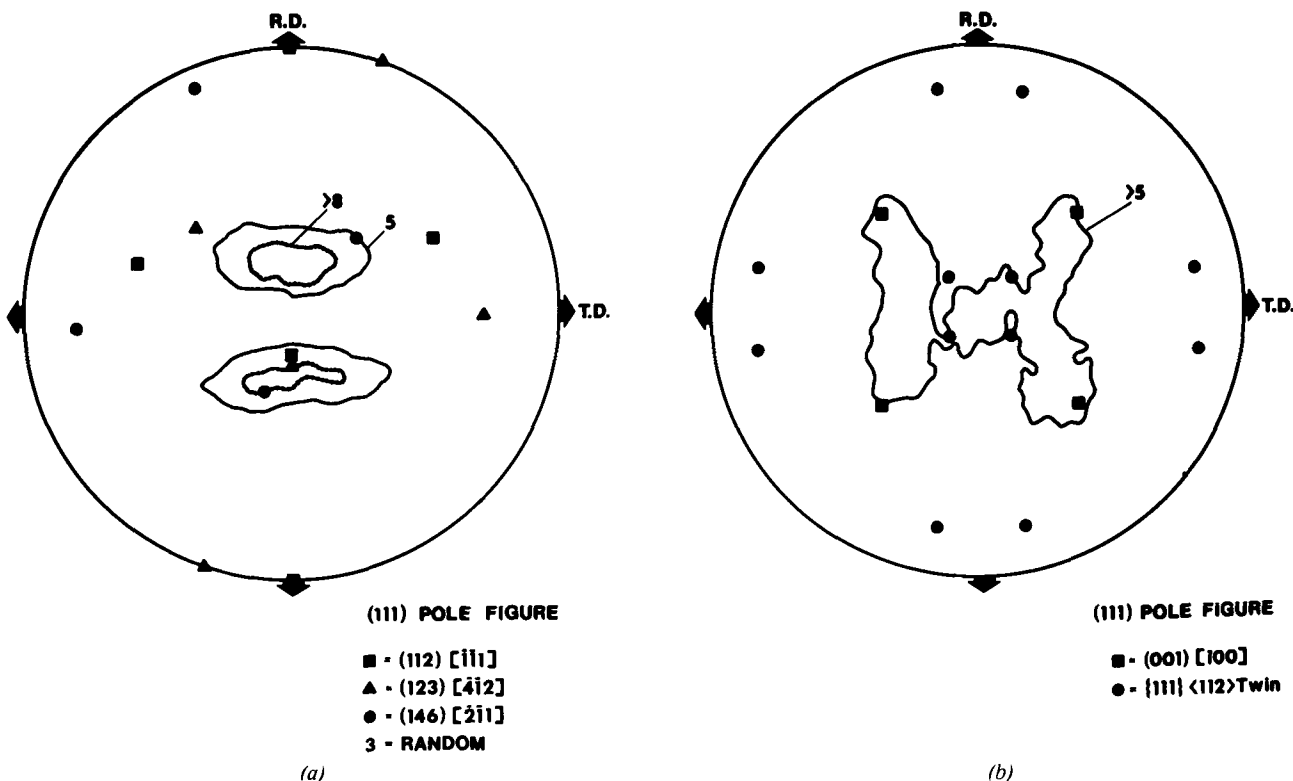

Fig. 4—(111) pole figures taken from AISI 304 steel. (a) Elongated austenite grains, (b) Equiaxed austenite grains.

highest pole density to be approximately 55 deg off the normal to the rolling plane and on great circles that make angles of 45 deg with the great circle connecting the normal to the rolling plane and the rolling direction.

A summary of the results presented above indicated quite clearly that the austenites in the two steels have exhibited very similar behavior during hot rolling.

The next area of investigation concerned the nature or origin of the equiaxed grains observed after high temperature rolling, e.g. Figs. 1(a) and 2(a). Of particular interest in this current study was the investigation of whether or not these new, equiaxed grains had formed as a result of dynamic recrystallization. One method that has been used to determine whether or not dynamic recrystallization had occurred involves observing the dislocation structure found in hot deformed and quenched specimens.[5, 17,18] Toward this end, a study was made of the dislocation arrangements that were present in four specimens of the 304 steel. The specimens investigated had been given the following treatments:

1. Hot rolled at 790 °C, WQ. This specimen exhibited all elongated grains.

2. Hot rolled at 1095 °C, WQ. This specimen exhibited all equiaxed grains.

3. Hot rolled at 1040 °C, WQ. This specimen exhibited a mixture of equiaxed and elongated grains.

4. Hot rolled at 1040 °C, annealed at 1120 °C for 30 min., WQ. This specimen contained all equiaxed grains.

The dislocation arrangements observed in the 304 steel that had been rolled at 790 were quite complex and inhomogeneous. Figure 5 shows typical examples of the dislocations observed; these micrographs illustrate the heterogeneous nature of the dislocation distribution. The observations made in the present study on material rolled at 790 °C are quite similar to several of the observations made in the investigation of cold rolled 304 steel by Blicharski and Gorczyca.[19] That investigation found that the deformation substructure consisted of subgrains, deformation bands and deformation twins. The subgrains observed in the present study, Fig. 5, were also very similar to the subgrain arrangements postulated previously to explain the spatial distribution of V(CN) precipitates observed in the V-steel after low temperature rolling and subsequent holding.[3] Figure 6 shows a typical distribution of V(CN) precipitates that had formed on the subgrain structure of austenite in specimens of V-steel that had been rolled and held at the temperatures 790 and 870 °C.[3]

The fully equiaxed, as-rolled grains possessed a more uniform dislocation substructure, where the dislocation density appeared to vary from being relatively low in the center portion of the grains to being relatively high in the vicinity of grain boundaries and twins. These dislocations are arranged in a diffuse, three-dimensional network that appears to be an early stage of subgrain formation. A typical example of the dislocation arrangements found within the equiaxed grains is given in Fig. 7.

The mixed grain structure found after rolling at in-

termediate temperatures exhibited both types of dislocation arrangements described above, sometimes even in adjacent grains as is shown in Fig. 8. The grain in the upper right portion of Fig. 8 shows a
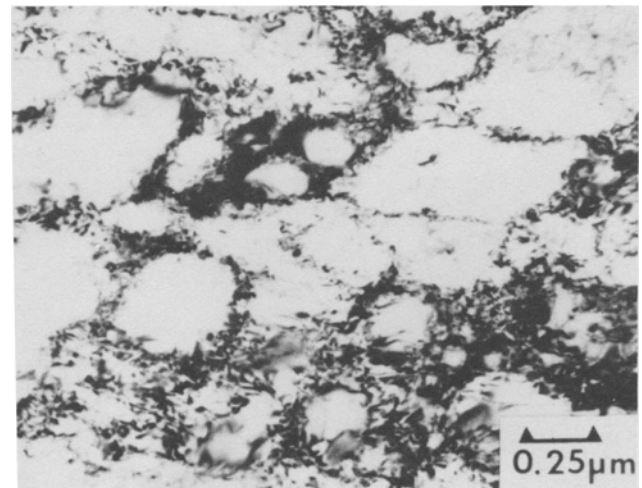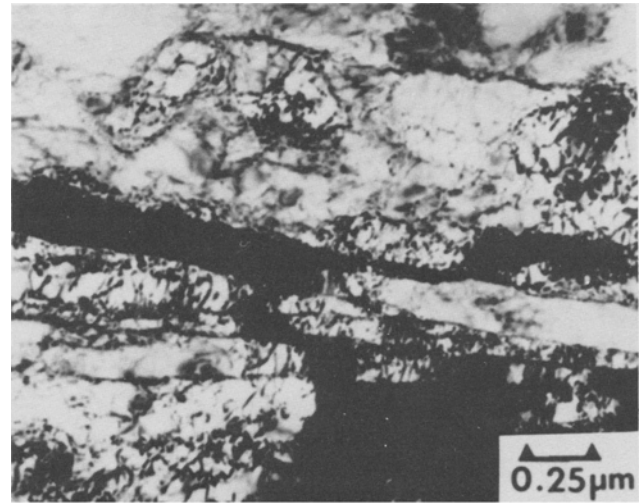
(b)

Fig. 5—Dislocation substructures in AISI 304 steel hot rolled at 790 °C. (a) Elongated subgrains, (b) Equiaxed subgrains.
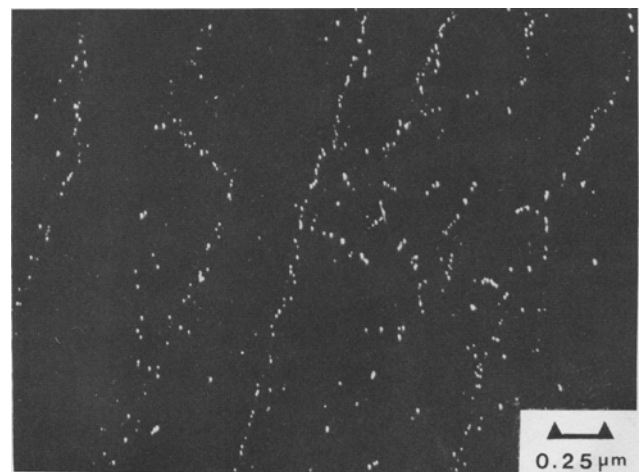

Fig. 6—Precipitation of V(CN) on subgrain boundaries of austenite in V-steel. Specimen rolled and held at 790 °C.
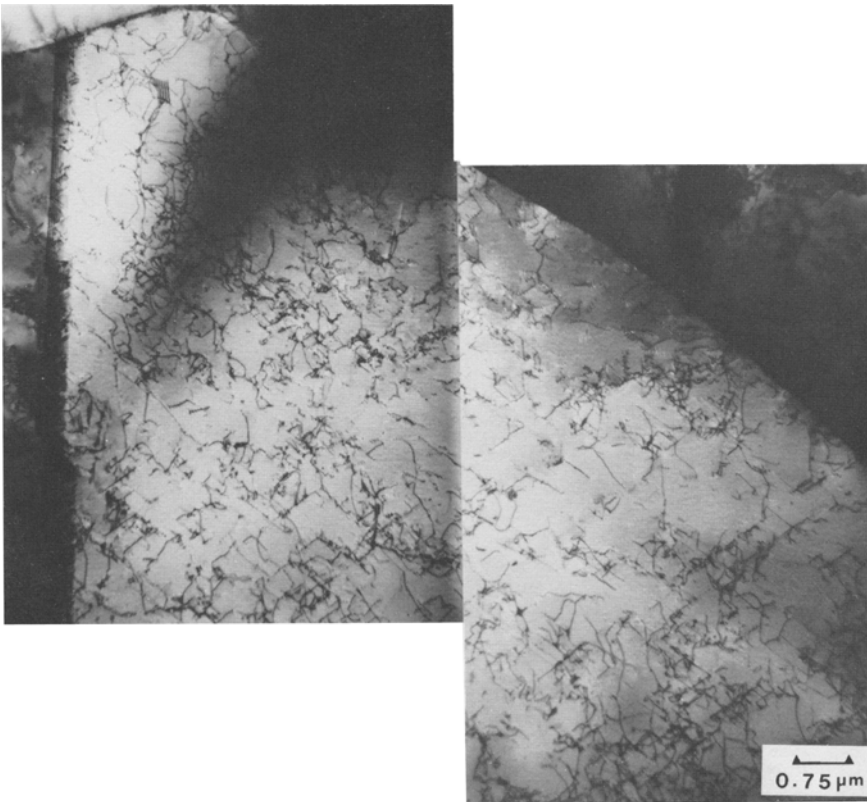
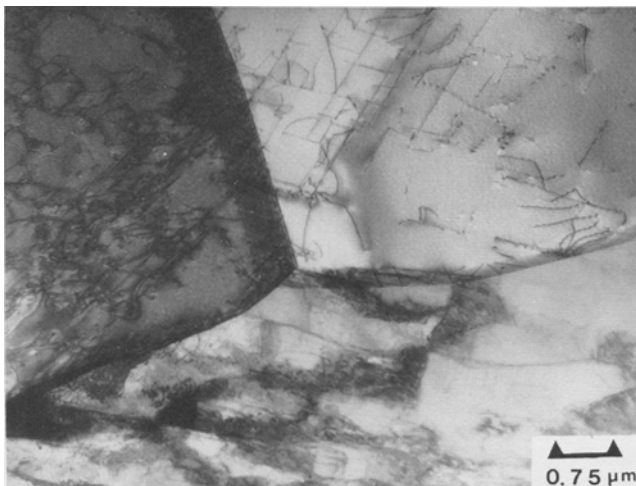Fig. 7—Dislocation substructures in AISI 304 steel hot rolled at 1095 °C.

Fig. 8—Spectrum of dislocation substructures found in different grains in AISI 304 steel hot rolled at 1040 °C.

dislocation arrangement of the type noted above in Fig. 7 for the as-rolled equiaxed grains. The grain at the bottom of Fig. 8 shows the same diffuse cell structure described above for the elongated as-rolled grains (Fig. 5).

The specimen that was held at 1120 °C subsequent to rolling at 1040 °C exhibited very few dislocations and virtually no substructure, but rather the clean, strain-free grains indicative of material that had undergone static recrystallization.

The differences in dislocation density expected for static and dynamic recrystallization would also be expected to be manifested in microhardness test results.

To test this hypothesis, two specimens were treated to exhibit the same approximate grain size. The grain size in one specimen was achieved by conventional cold rolling and annealing and in the second by hot rolling. The testing load used (25 g) was small enough to allow hardness indentations to be made within individual grains. The average hardness of the grains from the cold rolled and annealed specimen was KHN 151 whereas the grains from the hot rolled specimen was KHN 190. This microhardness study showed that the hot rolled material possessed a statistically significant higher hardness than the statically recrystallized material.

DISCUSSION

As indicated above, there were two major goals of this work: 1) To demonstrate that the V-steel and 304 stainless steel exhibit similar behavior during hot rolling, and 2) To investigate the origin of the equiaxed grains that result from high temperature hot rolling. These two topics will be discussed separately.

Based on the results of the optical metallography and texture studies, it would appear that the austenite in the two steels showed virtually the same behavior during hot rolling. That is, as the temperature of hot rolling was lowered, both steels exhibited a very similar variation in as-rolled grain morphology and texture. A comparison of the variation in grain morphology with rolling temperature for the prior-austenite in the V-steel (Fig. 1) and the austenite in the 304 steel (Fig. 2) shows this similarity in behavior. The texture study provided additional evidence sup-

porting the microstructural observations. The similar nature of the (110) pole figures for the V-steel and the (111) pole figures for the 304 steel in specimens that exhibited similar as-rolled austenite grain shapes indicated that similar metallurgical conditions prevailed in the two austenites.

The pole distribution shown in Figures 3a and 4a are very interesting since they correspond well to the most intense portions of the deformation texture (123) [412] + (146) [211] found by Hu and Goodman for 304 austenitic steel that was rolled at 800 °C.[20]* This tex-

*Although the pole figures shown in Fig. 3 and 4 are similar, in general, to those in the literature to which they are being compared, they are by no means identical. There are several reasons for this lack of detailed agreement. These include differences in initial grain size, deformation zone geometry, total rolling reduction, rolling technique (*i.e.* single *vs.* multiple, reverse passes) and the presence of concurrent deformation. These factors are either known[12,21,22] or may be expected to influence the final appearance of a given pole figure.

ture is the so-called "pure metal" or "copper-type" deformation texture.[12,20]

That the two steels would exhibit similar deformation textures is somewhat surprising in light of three presumably well-established facts: i) The large difference in alloy content contained in the two steels, ii) The influence of solute level on stacking fault energy of FCC alloys[23] and iii) The influence of alloy content or stacking fault energy on deformation texture.[12,22] Clearly, the results of the present investigation show that when the deformation occurs at elevated temperatures, the high temperature of deformation is able to offset any large variation in stacking fault energy that might exist due to differences in alloying. Similar effects have been shown in several previous studies in which an increase in the deformation temperature had caused a change of deformation texture from being "alloy" type to "pure metal" type.[20,24] Observations of this kind may help explain why the two steels in the present experiment have the same deformation texture, and also why no dissociated dislocations were found in any of the conditions observed in the 304 steel.

The pole distributions shown in Fig. 3(b) and 4(b) correspond reasonably well to the regions of high pole density expected for the ideal (001) [100] recrystallization texture found in 304 austenitic steel.[25] This ideal recrystallization or "cube" texture has been shown to result from the static recrystallization of the "copper-type" deformation texture (123) [412] + (146) [211].[25,26] The results of the present investigation appear to coincide quite well with these previously published.[25,26]

One point of clarification is required for each pole figure determined from specimens that exhibit the equiaxed austenite grain morphology, Figs. 3(b) and 4(b). First, the interpretation of Fig. 3(b) is complicated by the possibility that there can be two different ways by which this pole distribution may occur. The first is by the static recrystallization of austenite that exhibits the "copper-type" deformation texture. The second is by the deformation twinning of the ferrite or, more likely, martensite that had transformed from austenite which exhibited the "copper-type" deformation texture prior to transformation. For ex-

ample, if the austenite exhibits the "copper-type" deformation texture and is then quenched to martensite, and if this martensite happened to experience deformation twinning during its formation, the twinned portion of the inherited deformation texture could cause poles to be located in the vicinity of the cube texture. Hence, the twinning of the inherited deformation texture could possibly be mistaken for the cube texture. In the present study the pole distribution shown in Fig. 3(b) has not occurred by this twinning mechanism since a) no twins were observed in the low carbon martensite and b) other regions of high pole density would also be expected because of this twinning argument and these were not observed.

The second point of clarification must be made regarding the (111) pole figure of the 304 steel that exhibited equiaxed, as-rolled grains, Fig. 4(b). Twins are shown in Fig. 2(a). Evidence for twinning was also present in the (111) pole figure, Fig. 4(b). The pole locations expected from the twinning of the cube recrystallization texture are denoted by the filled circular symbols in Fig. 4(b).

Since the microstructure and texture results show that the two steels exhibit fairly similar behavior during hot rolling, the subsequent observations made on the 304 steel may be considered general and can be extended to elucidate the hot rolling behavior of low alloy steel austenite.

The second major goal of the present research was to try to identify the nature or origin of the equiaxed grains that resulted from high temperature rolling. The salient observations made in the present study were:

1. The as-rolled, equiaxed grains contained a rather high dislocation density. These dislocations were arranged in a three-dimensional network that could be interpreted as an early stage of cell or subgrain formation. The dislocation density in the vicinity of the grain boundaries was higher than in the bulk of the grains.

2. The as-rolled, equiaxed grains exhibited a (001) [100] texture which is indicative of static recrystallization in many FCC metals and alloys that have the "copper-type" deformation texture.

3. The metallurgical condition of the as-rolled, equiaxed grains is unstable relative to subsequent static annealing.

These three observations appear to favor dynamic recrystallization as the process that has led to the formation of the as-rolled, equiaxed grains in the 304 steel rolled at 1095 °C. Continuous recrystallization can be eliminated as a possible mechanism based on the texture results. If the equiaxed grains had formed via continuous recrystallization, the texture exhibited would have been a sharp deformation texture rather than the recrystallization texture that was actually observed, Figure 4(b).[27]

The results also appear to eliminate the possibility that the equiaxed grains had formed by static recrystallization after the deformation, *i.e.* during the quench. For example, the rather high dislocation density that was found throughout the equiaxed grains,

Fig. 7, is not consistent with the microstructure that would be expected if static recrystallization had occurred after deformation. Grains that form via static recrystallization, in the absence of concurrent deformation, contain a very low dislocation density, certainly one that is much lower than that exhibited by the equiaxed, as-rolled grains, Fig. 7.[4,28] This difference in dislocation density is also responsible for the difference in microhardness observed in the specimens that were hot rolled and cold rolled and annealed to the same grain size.

The same argument concerning dislocation densities that was used to eliminate static recrystallization subsequent to rolling can also be used to eliminate metadynamic recrystallization as a candidate mechanism for the formation of the equiaxed grains. Metadynamic recrystallization occurs by nucleation that takes place dynamically, *i.e.* during deformation, and by growth that takes place statically, *i.e.* after deformation. Since the growth takes place after deformation, the grains that form via metadynamic recrystallization would be expected to have a very low dislocation density, possibly as low as statically recrystallized grains. Once again, the high dislocation density observed in the as-rolled, equiaxed grains is not consistent with the internal microstructure expected from metadynamic recrystallization.

Two observations were made that are germane to the understanding of the nature or origin of the equiaxed grains that result from high temperature hot rolling. The first observation of importance is that the grains exhibited a standard FCC recrystallizaton texture (110) [100]. This cube-type texture is one of the standard recrystallization textures that has been frequently observed after conventional cold rolling and subsequent annealing. In these studies, the cube-type texture resulted from the static recrystallization of the copper-type deformation texture. The mechanism of static recrystallization is the motion of high mobility boundaries. The orientation difference that exists across these boundaries results in the deformation texture being transformed to the recrystallization texture as the boundary moves into the deformed region.[4,28,30] The presence of the cube-type recrystallization texture in the hot rolled materials suggests that the motion of highly mobile boundaries might be applicable in dynamic as well as static recrystallization.

The second observation of importance concerns the location of freshly nucleated dynamically recrystallized grains. There appeared to be two types of areas which could act as sites for the nucleation of dynamically recrystallized grains. These sites were a) the original grain boundaries and b) deformation bands and incoherent twin boundaries that ran through the original grains. When recrystallized grains nucleate either on or in the vicinity of grain boundaries during static recrystallization, the mechanism of nucleation is usually the strain-induced boundary migration model suggested by Bailey and Hirsch.[28-30] Unfortunately, the present study has not been able to assess whether or not any of these nucleation mechanisms were definitely operating during dynamic recrystallization; however, there is considerable circumstantial evidence indicating that they both may be operating.

In summary, the equiaxed grains that resulted from high temperature rolling have been shown to a) contain a rather high dislocation density and b) exhibit a cube-type recrystallization texture. These results have suggested that the mechanism most likely to be responsible for the formation of the equiaxed grains is dynamic recrystallization, since the other possible mechanisms of formation, *i.e.* continuous recrystallization, static recrystallization after deformation and metadynamic recrystallization can be eliminated as candidate mechanisms. The results of the present study have also led to a better understanding of dynamic recrystallization. The presence of the cube-type recrystallization texture in the equiaxed grains strongly implies that the motion of highly mobile boundaries plays a role in the formation of these grains. Thus, it would appear, that dynamic recrystallization combines both a deformation stage and a recrystallization stage where the motion of highly mobile boundaries occurs. The operation of these stages would explain: i) the presence of the cube-type recrystallization texture, ii) the weakness of this texture and iii) the presence of a dislocation density within the equiaxed grains that is much higher than that which is normally associated with static recrystallization.

CONCLUSIONS

The research described above has led to two major conclusions:

1. Despite the large differences in alloy content, AISI 304 austenitic steel and the low alloy V-steel exhibit similar microstructural behavior during hot rolling.

2. The equiaxed grains that form during high temperature hot rolling contain a rather high dislocation density, and, at the same time, exhibit one of the standard cube-type FCC recrystallization textures. These observations have been interpreted to mean that the grains had formed by dynamic recrystallization and that dynamic recrystallization may consist of processes that are similar to those associated with static recrystallization.

ACKNOWLEDGMENTS

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