Recrystallization of Austenite

L. J. CUDDY, J. J. BAUWIN, AND J. C. RALEY

A laboratory simulation of hot-rolling schedules, which employs a sequence of high-speed compressions performed over a range of temperatures, was used to determine the effects of composition and processing variables on the state of the austenite at several points in the hot-rolling procedure. The type and quantity of carbide and/or nitride forming elements (Nb(Cb), V, Al) in the steels were varied to produce precipitates at various points in the hot-working range of temperatures. Time-temperature-deformation schedules were designed around schedules currently used in the hot-reduction of plate steels; variations were introduced to determine the effects of processing variables on the development of the austenite structures. At various points in the deformation schedule, specimens were quenched out and changes in the austenite grain structure were examined. Changes in the volume and aspect ratio of the prior-austenite grains were used as measures of the degree of austenite recrystallization. At temperatures above the precipitation-start temperature, light reductions (< 10 pct) caused grain coarsening. Coarsening could be eliminated by increasing the reduction per pass, decreasing the rolling temperatures, or increasing the concentration of carbide or nitride-forming elements to raise the precipitation-start temperature. It is clear from these results that attempts to study structural changes that occur during plate rolling by using a few high-reduction passes may produce erroneous results.

INTRODUCTION

CONTROLLED rolling has proved effective in refining the ferrite grain size of HSLA steels. This results principally from control of the size and shape of the austenite grain structure at the point of transformation. Precipitation of alloy carbides and/or nitrides inhibits the growth of recrystallized austenite grains at high (roughing) temperatures, and at the lower finishing temperature, precipitates apparently inhibit recrystallization altogether. Thus, at transformation, fine, ribbon-shaped austenite grains are present. This state produces the maximum grain boundary area per unit volume, thereby producing the maximum density of sites for the nucleation of ferrite. Figure 1 shows how the surface-to-volume ratio (S/V) or grain boundary density of three initial grain sizes increases as the grains are compressed from spheres of radius b_0 to a pancake of height b (*i.e.*, as b/b_0 decreases). Each curve is for constant volume. If, during the compression, the flattened grain recrystallizes to smaller, equiaxed grains (jumps to a smaller b_0 curve) a further increase in S/V is realized. By following a path of repeated flattening (decrease in b/b_0) and recrystallization (decrease in b_0) and final flattening as illustrated in Fig. 1, the density of ferrite nucleation sites can be increased by two orders of magnitude.

To assure that the precipitation necessary to produce such refinement will occur in the hot-rolling range, one must design steels with a knowledge of the solution/precipitation behavior of the alloy carbides and nitrides. Figure 2 shows the equilibrium solubility products of several alloy carbides and nitrides in austenite[1,2] in the range of rolling temperatures and compositions commonly used for HSLA steels. These data indicate that alloy precipitates can form over most of the rolling range, but do not say that the kinetics of precipitation are fast enough to assure that the precipitates form and, if formed, whether they will inhibit grain boundary motion.

Evidence of particle-boundary interaction can be found in the results of grain coarsening experiments, such as those shown in Fig. 3.[3] Upon heating, the grain size of plain carbon steel (C) is already increasing rapidly at 900°C, whereas the grain sizes of the alloy steels (V, Al, Nb) remain fine to higher temperatures then increase rather abruptly and finally increase with further temperature increase at a rate comparable to that of the C steel. The temperature at which rapid coarsening occurs for each steel correlates roughly with the temperature at which, under equilibrium condition, the precipitate would be expected to be entirely dissolved (shown by the arrow on each curve).

Such data, however, tell us nothing of how the precipitates will affect boundary mobility during cooling with superimposed plastic work. That they are effective is clear from the success of fine-grain practice. From a considerable volume of mill trials and laboratory work, it has been found that the effectiveness of the precipitates is extremely sensitive to steel composition and soaking procedure, as would be expected from Fig. 2, and to the time-temperature-deformation schedule.[4-6] Of these factors, the last is the most difficult to study in detail because of the impossibility of extracting samples during hot rolling of slabs and plates. Consequently, laboratory rolling and simulations have been employed.[7-9] Generally, these have been limited to a few, high-reduction passes or to modes of deformation that do not approximate hot rolling. Recently, we have developed a procedure

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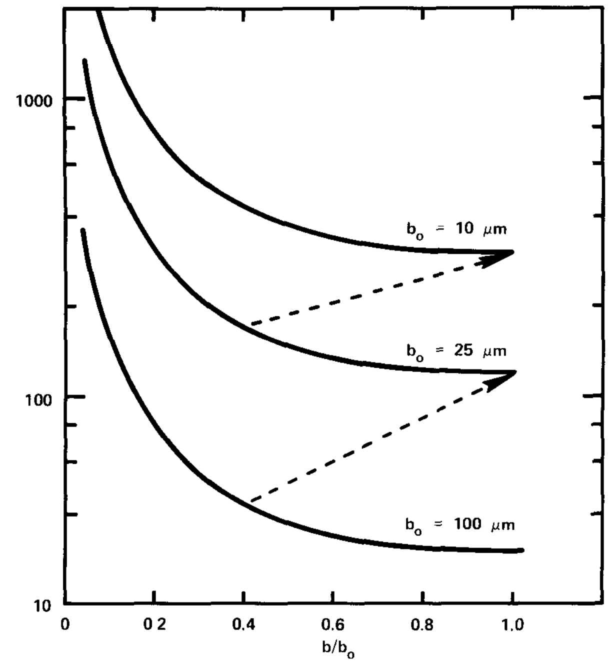

Fig. 1—Effects of original grain diameter (b_0) and reduction in height (b/b_0) on grain boundary area per unit volume (S/V).

which imposes on the specimen a sequence of high-rate compressions during controlled cooling at rates that simulate that of steel plates. With this technique, we can duplicate the time-temperature-reduction schedules used in the mills while extracting specimens at any point in the schedule to examine the structure of the austenite. This technique is being employed to examine the development of the structure of austenites of several compositions during simulated hot rolling between 1150 and 750°C (2100 to 1380°F). This paper discusses results of simulations of the roughing passes only; that is, deformation in the range 1150 to 930°C (2100 to 1700°F).

MATERIALS AND EXPERIMENTAL WORK

The compositions of the five steels used in this study are shown in Table I, together with the precipitates to be expected in each and the maximum temperatures T_c at which precipitation should begin under equilibrium conditions. In addition to designing these steels for their precipitation effects, there was added to each 1.4 wt pct Mn to assure sufficient hardenability so that the prior austenite grain structure could be observed in the deformed specimens after quenching.

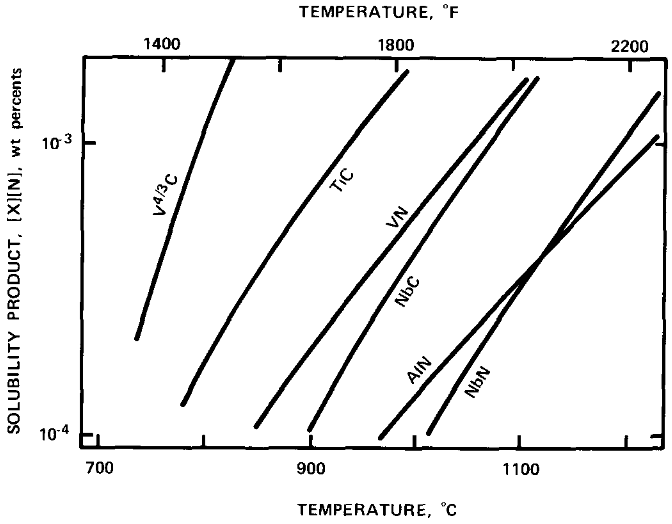

Fig. 2—Solubility products for alloy carbides and nitrides in austenite.[1,2]

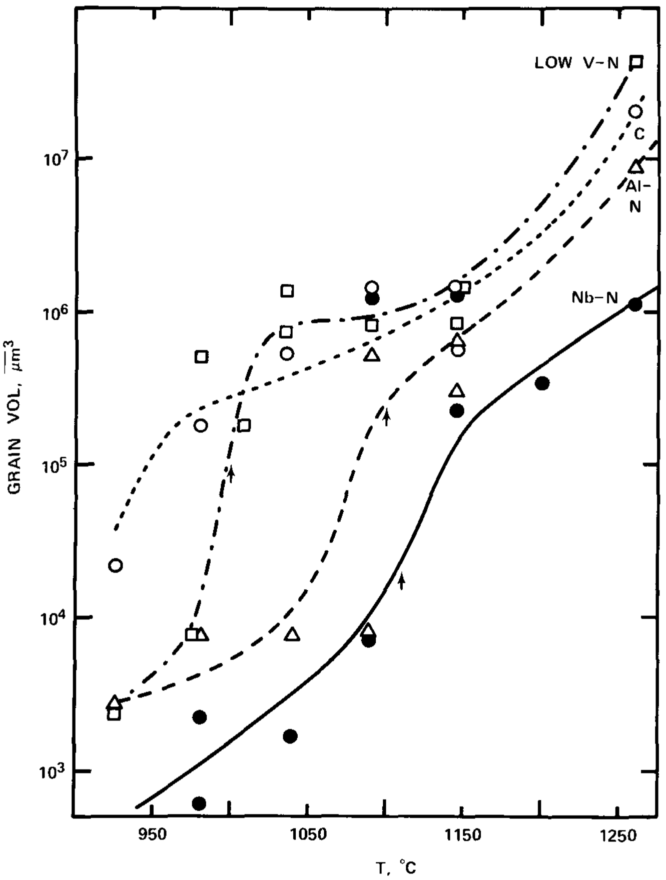

Fig. 3—Grain coarsening of four steels held 15 to 30 min at each temperature.

Cylindrical compression specimens 10 mm in diam by 19 mm high were mounted between anvils in an MTS machine, as shown in Fig. 4. A shielded thermocouple inserted into the specimen center was used to control the temperature of the radiant furnace that surrounded the specimen and anvils. The specimen was soaked at 1150 or 1200°C for 30 min then control-cooled at the plate cooling rate by means of Data Trak control of the radiant furnace. At preselected temperatures, a sequence of high-strain-rate compressions ($\dot{\epsilon}$ = 10 to 30 s^{-1}) was imposed on the specimen to simulate rolling passes. Table II and Fig. 5 show typi-

Table I. Steel Compositions, the Precipitate, and the Maximum Temperature at which Precipitate Can Form under Equilibrium Conditions

Steel	C	Mn	Si	N	Other	Precipitate	T_c, °C
Plain C	0.065	1.37	0.27	0.007	–	–	–
Al-N	0.065	1.43	0.26	0.005	0.068 Al	AlN	1100
Low V-N	0.078	1.38	0.27	0.006	0.098 V	VN	1000
High V-N	0.010	1.40	0.25	0.015	0.250 V	VN	1200
Nb-N	0.041	1.42	0.24	0.009	0.090 Nb	NbN	1150
						NbC	1220

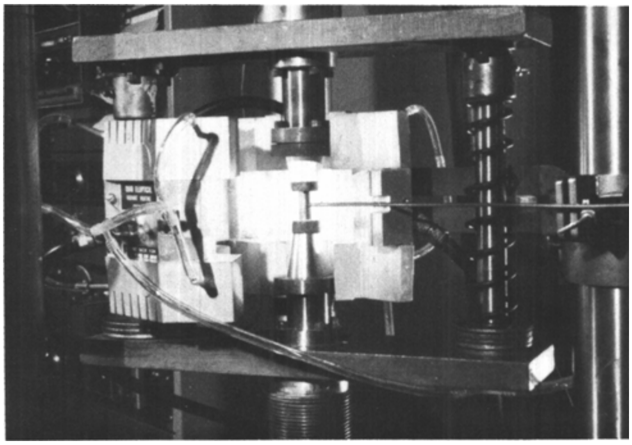

Fig. 4—Specimen (center) in hot compression apparatus.

cal time/temperature/deformation schedules used in this simulation. Figure 6 shows force-displacement data recorded on an oscilloscope during the deformation sequence.

Specimens were withdrawn at any preselected point in the deformation schedule and immediately quenched in iced brine. The most satisfactory procedures for revealing the austenite grain structure, particularly in highly deformed specimens, consisted of either immersion for 1 to 2 min in saturated aqueous picral at 80°C, or immersion for about 5 min in boiling saturated aqueous picral modified with a wetting agent (sodium tridecylbenzene sulfonate).

Average grain dimensions parallel to (h) and normal to (l) the compression axis were determined by linear intercept. From these measurements, the average grain volume l^2h and the grain aspect ratio l/h were calculated assuming axial symmetry of the grains. These parameters were taken as indicators of the degree of recrystallization or work in the grains. For example, at one extreme, a decrease in volume coupled with an l/h = 1.0 indicates complete recrystallization; constant volume with an increase of l/h indicates the accumulation of work in unrecrystallized grains.

Table II. Time-Temperature-Deformation Schedules

Pass	Time, s	Temperature, °C	t/t_0	Reduction per Pass, Pct
A. Four-Pass, High-Reheat Schedule. Total Reduction = 71 Pct				
0	0	1260	1.00	–
1	12	1180	0.66	34
2	50	1100	0.46	30
3	80	1010	0.34	26
4	140	925	0.29	16
B. Twelve-Pass, Low-Reheat Schedule. Total Reduction = 55 Pct				
0	0	1150	1.00	–
1	12	1140	0.92	8
2	18	1135	0.89	3
3	82	1090	0.85	4
4	96	1080	0.82	4
5	128	1060	0.77	6
6	144	1050	0.73	5
7	205	1025	0.68	7
8	228	1010	0.64	6
9	262	990	0.57	11
10	300	970	0.54	5
11	327	950	0.49	9
12	360	930	0.45	8

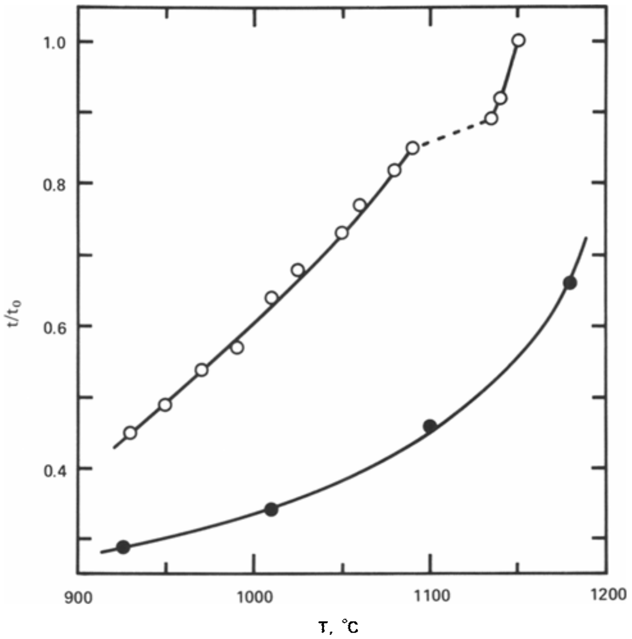

Fig. 5—Temperature-reduction schedules for compression of a cylinder from original height t_0 to t. ○ Twelve-pass schedule, ● Four-pass schedule, Table II.

RESULTS AND DISCUSSION

The reduction per pass and the solubility product of the alloy nitride at each reduction turn out to be the two important factors in control of austenite grain structure during deformation at the high temperature end of the rolling schedule.

This conclusion derives from a comparison of results of the grain structures in several alloy steels deformed according to the two schedules shown in Table II. As shown in the micrographs, Fig. 7, and in the data, Fig. 8, each pass of the four-pass schedule

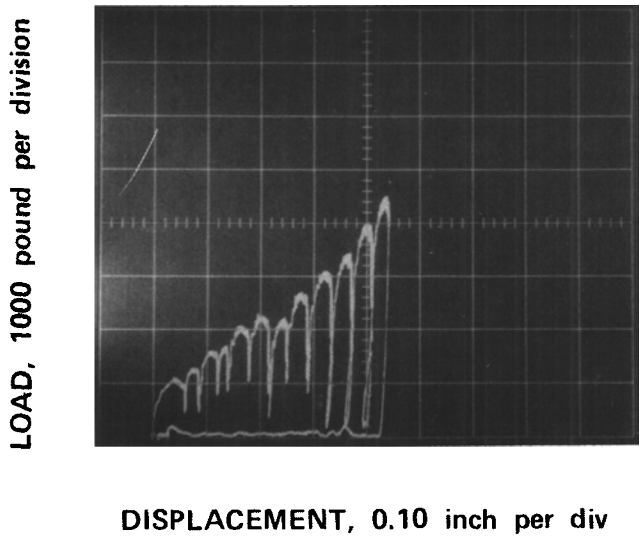

Fig. 6—Load displacement data recorded during the 12-step reduction of the Nb steel between 1150 and 930°C. Total reduction 0.52 pct.

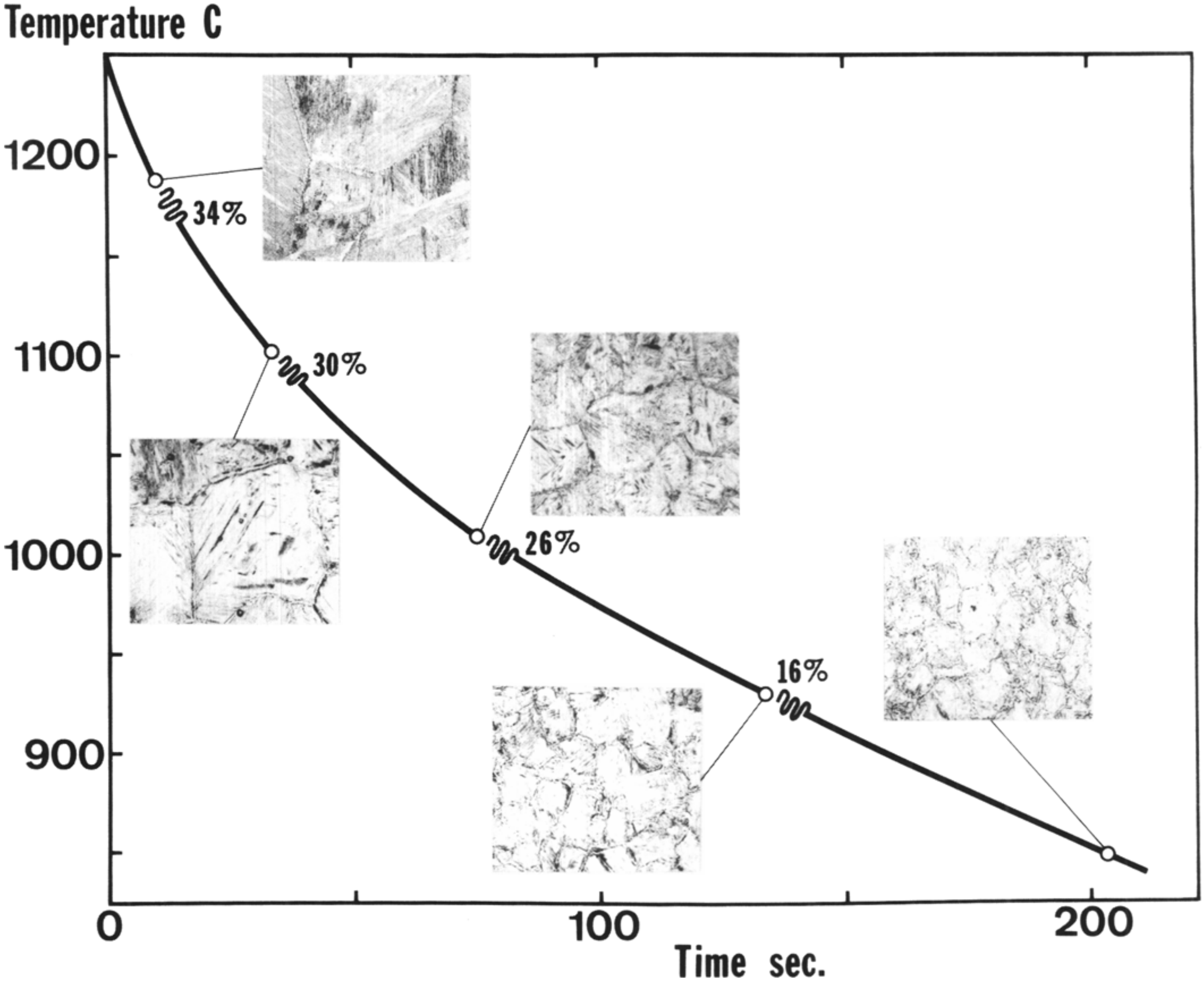

Fig. 7—Austenite grain structures developed in the Nb steel during the 4 pass reduction Magnification 156 times.

produced a refinement in austenite grain size. Also, after each pass, the grain aspect ratio (not shown) remained 1.0. These results indicate that full recrystallization occurred in all the steels after each pass. Compare these results with the grain structures developed in a simulation that closely approximates a twelve-pass roughing schedule in the same temperature range, shown in Fig. 9. Despite the 30-min soak at 1150°C to achieve a stable grain size prior to deformation, the first few reductions produce grain growth rather than refinement in the C and low-V steels and a plateau in the Nb and Al steels. Below 1075°C, refinement occurs by recrystallization. This produces roughly equiaxed grains in all the steels except the Nb steel in which the grain aspect ratio rose to a value greater than 2.0.

The effect of reduction per pass is shown more clearly for the low V-N steel in Fig. 10. T_C on the diagram indicates the temperature at which precipitation of V-N could begin in this steel under equilibrium conditions. Clearly, grain coarsening occurs in the range where alloy nitrides are entirely in solution when light reductions are used. With heavy reductions, refinement occurs even in the range of complete solubility of the alloy nitride.

The effect of the concentration of the alloy nitride-forming elements is demonstrated in Fig. 11 for the high and low V-N steels subjected to light reductions. As the concentration increases, the tendency toward growth after light reductions decreases. The arrows designate the critical temperature at which precipitation can begin. The grains in the low V-N steel grow when that steel is worked lightly above its critical temperature, whereas the high V-N steel which is worked entirely below its critical temperature shows no growth. Similarly, in Fig. 9, because they are worked entirely at or below their critical temperatures (arrows), the Nb and Al steels in Fig. 9 show no growth, whereas the low V-N and plain carbon steels which are worked above their critical temperatures, show initial grain coarsening.

Figure 12 demonstrates the relation between the critical precipitation-solution temperature, T_C and the temperatures at which rapid grain coarsening occurs in heating and refinement occurs during deformation/cooling for the low V-N steel. The critical

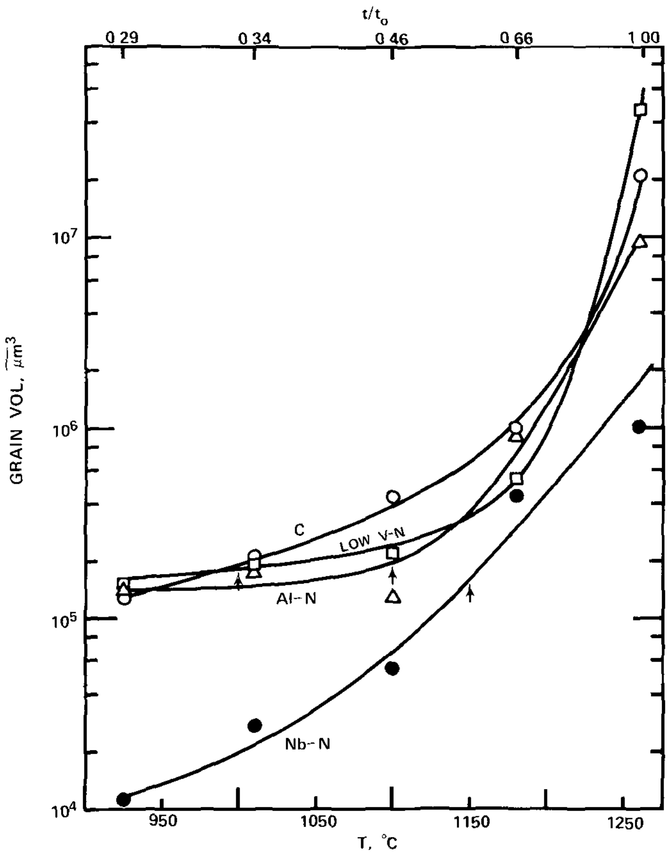

Fig. 8—Decrease in austenite grain size during a 4-pass hot deformation schedule. Total reduction 71 pct.

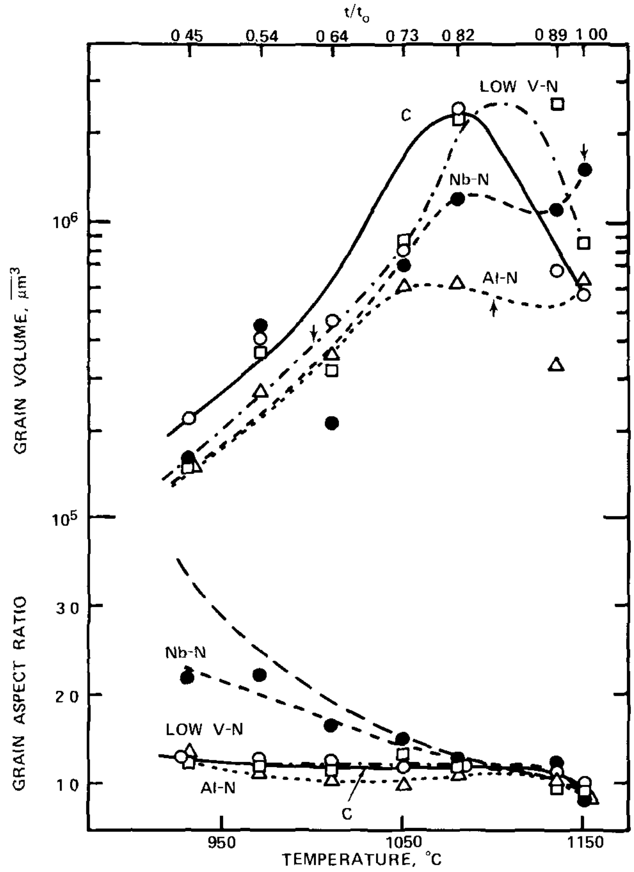

Fig. 9—Changes in austenite grain volume and grain aspect ratio during a 12-pass, hot-deformation schedule. Total reduction 55 pct.

temperature calculated from solubility data is a fair guide to the soaking temperature above which coarsening will occur and the deformation temperature below which a blow-up of grains can be avoided during light working. Concerning the latter, it was generally observed, as shown in Fig. 12, that refinement begins during deformation somewhat above the calculated critical temperature. This probably indicates that solute clusters, or fine precipitates that are not detected in solubility determinations impede boundary motion at these higher temperatures.

CONCLUSIONS

1. The deformation schedule and composition must be adjusted to avoid light reductions above the range of temperatures at which precipitation of alloy ni-

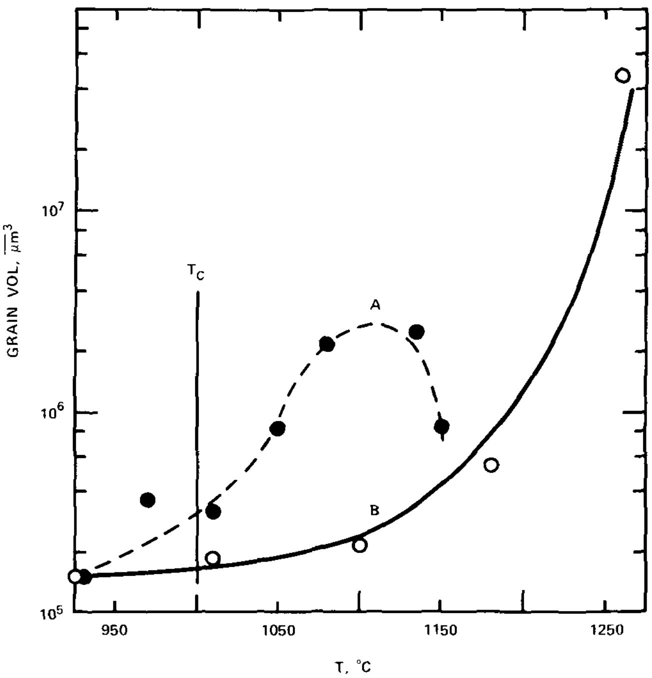

Fig. 10—Austenite grain volumes developed in the low V-N steel in the 12-pass (A) and 4-pass (B) deformation schedules.

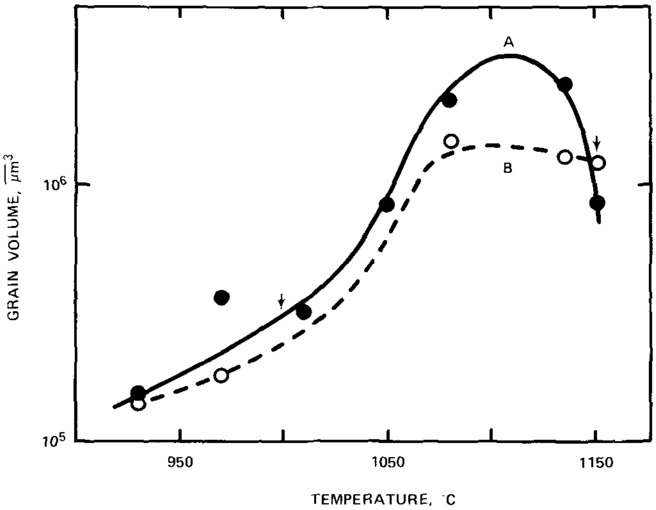

Fig. 11—Austenite grain volumes developed in the low V-N (A) and high V-N (B) steels during 12-pass reduction of 55 pct.

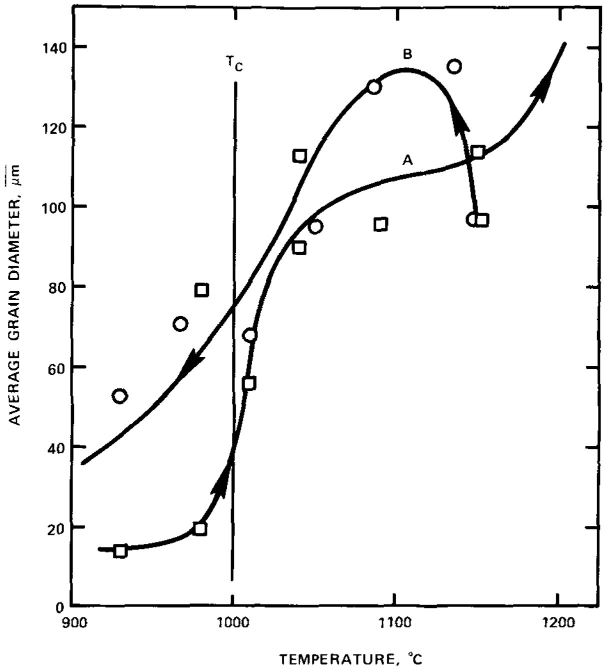

Fig. 12—Austenite grain size developed in the low V-N steel (A) during heating, and (B) during deformation while cooling.

trides and/or carbides begins. If the soaking temperature cannot be reduced or the draft increased, higher concentrations of nitride and/or carbide forming elements can be used. Generally, refinement of the austenite grains by recrystallization begins somewhat above the precipitation-start temperature determined from solubility-product data.

2. Strain can be accumulated (indicated by an increase in grain aspect ratios) only below the precipitation-start temperature. Even in this range, however, small accumulations can lead to recrystallization leading to finer equiaxed grains.

3. Rolling simulations using a few, high-reduction passes may not accurately indicate the manner in which the austenite structure develops in plate rolling schedules comprising numerous, light passes.

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