The Origin and Significance of Banding in 18Ni (250) Maraging Steel

P. H. Salmon Cox, B. G. Reisdorf, and G. E. Pellissier

Banding that occurred in plates rolled from the early production heats of 18Ni(250) maraging steel is described and related to the segregation of certain alloying elements (nickel, molybdenum, titanium), the extent of which was quantitatively evaluated by means of electron-microprobe analysis. The effect of banding on mechanical properties is discussed, with particular reference to observed directional differences in plane-strain fracture toughness of plates. It is shown that banding originates as interdendritic segregation during ingot solidification and persists in some degree through normal soaking and hot reduction to plate. The results of the study showed that heating sections of small laboratory-cast ingots at 2200°F for 4 hr was sufficient to markedly reduce microsegregation and to considerably improve mechanical properties. Hot rolling of 7-in.-thick ingot sections to ½-in.-thick plate effected a similar reduction of microsegregation, but resulted in even greater increases in ductility and toughness than that obtained by homogenization treatment alone.

DURING the past few years, considerable attention has been directed towards the low-carbon, high-alloy maraging steels and in particular towards the 18Ni-8Co-5Mo-0.4Ti alloy. The steels of this group, having an excellent combination of high strength and toughness, have a number of advantages over their more conventional medium-carbon low-alloy, quenched-and-tempered counterparts. In the annealed condition, the maraging steels are in the form of a ductile martensite; aging at a relatively low temperature, typically 900°F for 3 hr, increases greatly the strength through the precipitation of intermetallic compounds.

One problem in the early production heats of maraging steel was that the finished plate frequently displayed a banded structure. Previous work on other steels[1-6] had established that banding in wrought products is either a direct or an indirect consequence of chemical segregation, which occurs during solidification and persists to some extent through normal thermal and mechanical treatments. For example, Smith and others,[3] in a study of low-alloy steel, were able to correlate the severity of banding in the wrought product with the degree of interdendritic segregation of nickel and chromium in the as-cast ingot.

The effect of banding on the mechanical properties of steels is usually considered to be detrimental, although there is only limited evidence to suggest that a marked improvement in properties can be obtained with less heterogeneous structures. Comparison of the longitudinal and transverse tensile properties of banded and of homogenized 4340 steel showed that only the transverse ductility was improved by homogenization, but even then the improvement was not commercially significant.[4] Conversely, homogenization of through-the-thickness tension specimens of quenched-and-tempered steel plate, containing 1.47 pct Mn, increased the strength by as much as 10 pct and the tensile ductility by at least a factor of two.[5] This improvement was related to the elimination of manganese-rich bands, which also are one of the factors responsible for cold cracking in the heat-affected zone of metal-arc welds.[7]

P. H. SALMON COX, B. G. REISDORF, and G. E. PELLISSIER are Senior Research Engineer, Associate Research Consultant, and Senior Research Consultant—Physical Metallurgy, respectively, United States Steel Corp., Applied Research Laboratory, Monroeville, Pa.

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In the present study the nature and severity of banding in early commercial 18Ni(250) maraging steel plate and in laboratory-melted 18Ni(250) maraging steel plate was determined. The effects of banding on plane-strain fracture toughness and the effects of thermal homogenization treatments on the strength, tensile ductility, and toughness of 18Ni(250) maraging-steel as-cast ingots and rolled plate were evaluated. In addition, the effects of hot deformation by rolling on the mechanical properties of ingots were determined.

1) STUDIES OF BANDING IN EARLY PRODUCTION PLATE

The chemical composition of the steel (A) used in this part of the investigation is shown in Table I. Banding was not clearly evident in either as-rolled or annealed* plate, but annealed and aged** plate had a banded structure. The typical banded condition, Fig. 1, consists of layers of unetched austenite (white) and dark-etching martensite in a light-etching martensitic matrix. X-ray diffraction measurements showed that this steel contained more than 6 pct austenite.

*1500°F for 1 hr, air-cooled.
†1500°F for 1 hr, air-cooled; 900°F for 3 hrs, air-cooled.

An electron-probe X-ray microanalyzer (using a focused beam of electrons) was used to determine the composition of the bands and of the material between the bands with respect to the main alloying elements—nickel, molybdenum, titanium, and cobalt. The recorded X-ray intensities were converted to concentration values with the use of a standard of similar composition. To facilitate probe positioning, all analyses were conducted on specimens that had been given a light etch. The influence of this etching on the analytical results was negligible; analyses made on the identical area before and after etching yielded essentially the same concentration values.

The results of the electron-microprobe analyses at selected points revealed that the layers of austenite and adjacent dark-etching martensite contained greater amounts of nickel, molybdenum, and titanium than did the surrounding matrix, Table II. The austenite layers

Table I. Chemical Composition of 18Ni(250) Maraging Steels Investigated

Steel	Composition, pct C	Mn	P	S	Si	Ni	Mo	Co	Ti	Al
A	0.014	0.028	0.007	0.012	0.077	18.0	4.72	7.85	0.44	0.038
B	0.013	0.083	0.002	0.006	0.069	18.1	4.60	7.71	0.47	0.060
C	0.020	0.25	0.002	0.005	0.070	18.1	5.85	8.02	0.33	0.005
D	0.014	0.056	<0.001	0.006	0.049	18.0	4.96	8.08	0.42	0.013
E	0.012	0.053	0.001	0.006	0.017	17.9	4.96	8.08	0.40	0.006

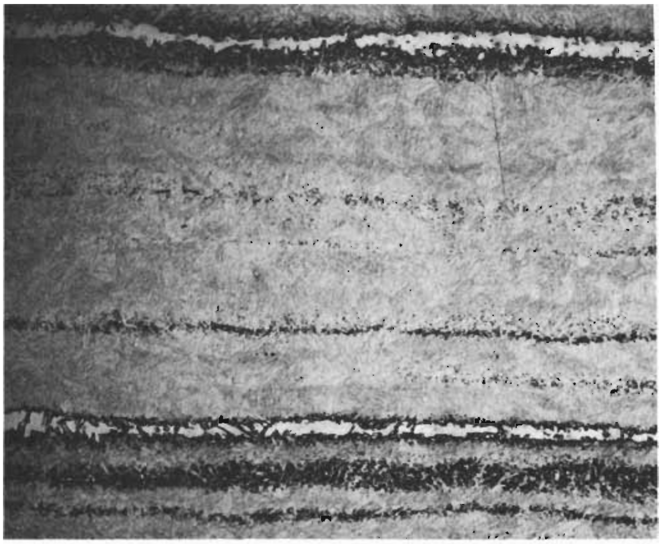


Fig. 1—Light micrograph of polished and etched longitudinal section showing segregation bands in rolled plate of maraging (250) steel in the annealed and aged condition. Magnification 185 times.

were richest in alloying element content, containing an average of 20.2 pct Ni, 6.3 pct Mo, and 0.81 pct Ti. The average concentrations for the light-etching martensite were 17.6 pct Ni, 4.3 pct Mo, and 0.38 pct Ti, representing differences in alloying element content of 2.6 pct Ni, 2.0 pct Mo, and 0.43 pct Ti from that of the austenite. The analyses included determinations of cobalt in the different bands, but no evidence of segregation of this element could be found; the recorded difference in cobalt concentration between austenite

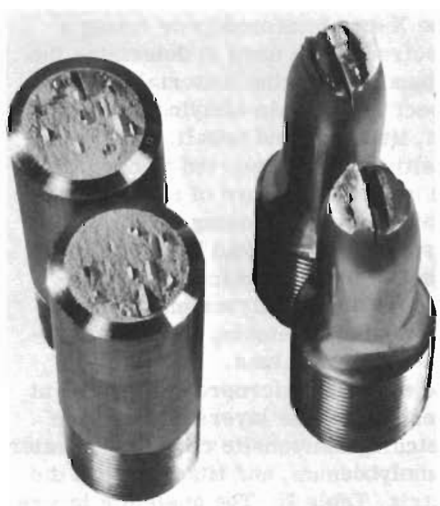


Fig. 2—Fracture appearance of smooth and notched tension-test specimens of maraging (250) steel showing longitudinal "splits" and "internal shear lips".

Table II. Electron-Probe X-Ray Microanalyses* of an Early Production Plate ($\frac{3}{4}$-in.-thick) 18Ni(250) Maraging Steel (A)

Area Analyzed	Element Concentration, pct Ni	Mo	Ti
Austenite bands (white)	20.2	6.3	0.81
Dark bands	19.4	5.0	0.56
Light bands (matrix)	17.6	4.3	0.38

*Average of three analyses.

and light-etching martensite was less than 0.1 pct, which was not significant. Although it has been shown that the dark-etching martensite contained greater amounts of nickel, molybdenum, and titanium than did the light-etching martensite, the cause of the dark-etching appearance of this high-alloy martensite is not known.

The fracture appearance of the mechanical-property test specimens was directly associated with the alloy-rich bands. Longitudinal "splits" were evident in the fracture surfaces of smooth tension-test specimens and "internal shear lips" were prevalent in the fracture surface of fatigue-cracked notched-round specimens that were used to determine plane-strain fracture toughness, Fig. 2, and in the fracture surface of Charpy V-notch (CVN) specimens. On polished-and-etched longitudinal sections (parallel to the rolling direction and perpendicular to the rolling plane), cut from the fractured specimens, the splits and internal shear lips intruded a considerable distance below the fracture surface along the alloy-rich bands, Fig. 3.

It is evident from the probe analyses that banding in 18Ni maraging steel is directly due to chemical segregation of nickel, molybdenum, and titanium; it will be shown later that this segregation is interdendritic and represents the last alloy-rich metal to freeze. On rolling, the pattern of segregation is changed and the interdendritic areas are flattened in the rolling plane and direction to give a layered or a banded structure. Furthermore, experiments to be described later in this paper show that this segregation can be markedly reduced by simple homogenization and rolling treatments.

2) EFFECT OF BANDING ON FRACTURE TOUGHNESS

To investigate the possibility that banding could produce directional differences in plane strain fracture toughness (K_{I_c}), a $1\frac{1}{8}$-in.-thick plate of 18Ni(250) maraging steel was rolled from a 500-lb ingot of 5- by 12-in. cross section that was teemed from a lab-

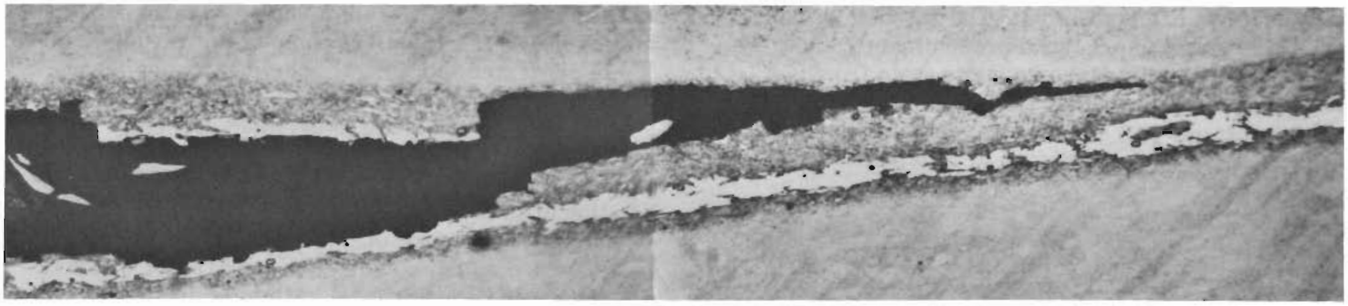


Fig. 3—Longitudinal section of a fractured smooth tension-test specimen showing the path of a "split" in relation to the alloy-rich band.

oratory heat. The chemical composition of this steel (B) is presented in Table I. Triplicate single-edge-notch (SEN), fatigue-cracked tension-test specimens were cut from the $1\frac{1}{8}$-in.-thick plate in four principal orientations, as shown in Fig. 4. The specimens were 0.14 in. thick by 1 in. wide by $4\frac{1}{2}$ in. long. The 45-deg V-notch plus fatigue-crack extension (produced in cyclic tension) was about $\frac{1}{3}$ in. deep. The specimens were eccentrically pin-loaded, with the pin-loading holes located $\frac{1}{3}$ in. from the notched edge. The distance between the loading points was $3\frac{1}{2}$ in. The crack "pop-in" (the transient instability due to sudden extension in opening mode) was detected with a microformer extensometer placed on the specimen. K_{I_c} values were computed by reference to a previously established curve of crack length vs compliance obtained by means of a clip-on compliance gage. To obtain adequate specimen length in the "D" specimens, extensions of the same material were carefully welded to each end of the initial 1-in.-long coupons. The specimens were tested in the annealed and aged condition.

The test results, Fig. 4, clearly demonstrated that, when the crack plane was located through the thickness of the plate and the crack propagation direction was either perpendicular to the rolling direction (A) or parallel to the rolling direction (B), the fracture toughness was significantly less (about 10 pct) than when the crack plane and direction were oriented normal to the plane of rolling (C). The lowest fracture toughness was obtained when both the crack plane and crack direction were parallel to the rolling plane (D); in this latter instance, the fracture toughness was about 20 pct lower than that of the longitudinal direction (A).

A metallographic examination of the fracture pro-

ORIENTATION	K_{I_c}, psi$\sqrt{in.}$ x 10^{-3}
A	80
B	77
C	90
D	63

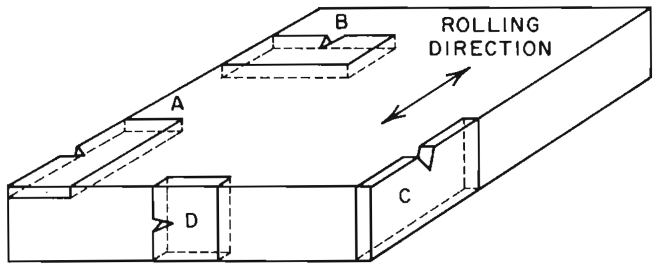
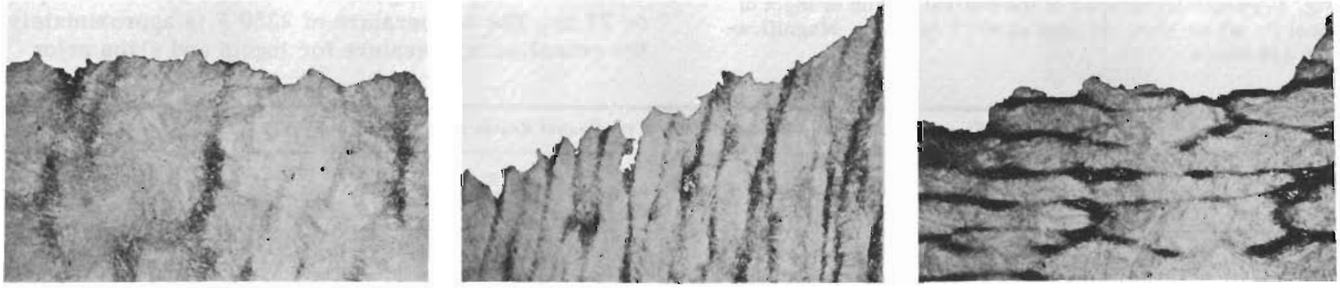


Fig. 4—Variations in fracture toughness (K_{I_c}) with specimen orientation in maraging (250) steel plate ($1\frac{1}{8}$ in. thick).

file of a polished and etched section of the A-oriented specimen (and B-oriented specimen) in the region of crack "pop-in" revealed that banding did not produce a marked influence on the path of fracture, Fig. 5(a); however, in the C-oriented specimen, Fig. 5(b), the fracture frequently detoured abruptly from its main course when it encountered segregation bands located perpendicular to the crack plane and direction. Thus segregation bands appeared to act as small internal crack arrestors, which tended to impede and divert the main crack, and forced it to follow a less direct path; this resulted in an increase in the measured fracture toughness of the material for this particu-



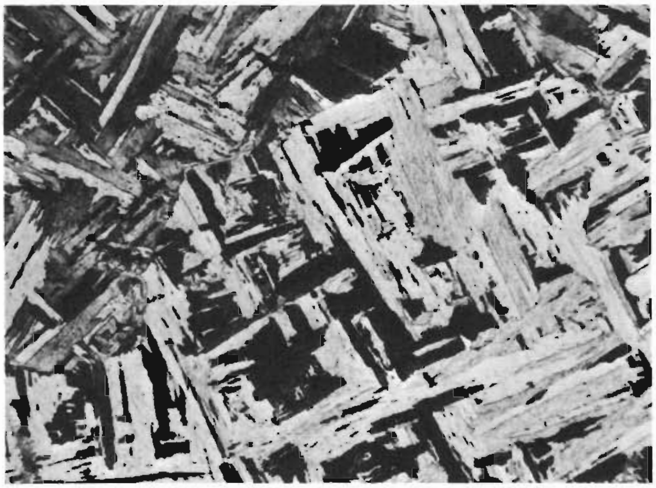
(a) (b) (c)

Fig. 5—Light micrographs of polished and etched sections showing fracture profile of "pop-in" region of different fracture-toughness test specimens: (a) specimen A; (b) specimen C; (c) specimen D. Magnification 120 times.

lar specimen orientation (C). The fracture of the D-oriented specimen occurred predominantly along the bands, occasionally stepping up or down from one band to another, Fig. 5(c). This type of fracture involved considerably less plastic deformation and energy absorption, and resulted in considerably lower values of K_{I_c}.

3) INTERDENDRITIC SEGREGATION IN A LABORATORY-CAST INGOT

To investigate interdendritic segregation in 18Ni maraging steel, a 100-lb 3- by 8- by 14-in. ingot was cast into a chill mold from a 300-lb vacuum-induction-melted heat. The chemical composition of this steel (C) is presented in Table I. Specimens were taken from the center of the ingot such that the plane of the surface to be examined was oriented perpendicular to the direction of heat extraction, and the dendritic structure was revealed by etching in ferric chloride solution after aging.



(a)

(b)

Fig. 6—Photomicrographs of the central region of ingot of Steel C: (a) as-cast; (b) aged at 900°F for 3 hr. Magnification 140 times.

The microstructure of the central region of this ingot, before aging, consisted of arrays of "massive" martensite plates, Fig. 6(a). After the ingot was aged, the center of the dendrites and the secondary and tertiary arms consisted of light-etching martensite whereas the interdendritic spaces and pools contained austenite surrounded by darker etching martensite, Fig. 6(b). Electron-probe X-ray microanalyses showed that the interdendritic areas (austenite) were alloy-rich and that the cores of the dendrites were depleted in alloying elements relative to the bulk composition. The differences between the average values of maximum concentration (C_{max}) and minimum concentration (C_{min}) at the center of the ingot were 2.5 pct Ni, 5.0 pct Mo, and 0.81 pct Ti, Table III; only small differences (0.2 pct) were detected in the concentrations of cobalt. The segregation ratios (the ratio of C_{max} to C_{min}) for nickel, molybdenum, and titanium were 1.14, 2.1, and 7.2, respectively. This represents a severely segregated structure in which the last-freezing liquid, located between the dendrite cores and branches, became progressively enriched in nickel, titanium, and molybdenum. The increased concentration of alloying elements in the interdendritic areas appears to affect the austenite-reversion temperature. Floreen and Decker[8] observed that austenite reversion could occur on heating to temperatures as low as 1000°F for times as short as 30 min. Evidently the austenite-reversion temperature of the interdendritic regions of Steel C was lowered so that reversion occurred during the standard aging treatment (900°F for 3 hr). Thus, the quantity of austenite in the aged microstructure provides a semiquantitative, indirect means for assessing the severity of segregation in the 18Ni maraging steels.

4) EFFECT OF HOMOGENIZATION ON INTERDENDRITIC SEGREGATION IN A LABORATORY-CAST INGOT

Specimens taken from the center of the ingot of Steel C, described in the previous section, were placed in a sealed silica tube to prevent oxidation, and then were thermally homogenized at 2250°F for either 4, 8, or 24 hr. The temperature of 2250°F is approximately the reheating temperature for ingots and slabs prior

Table III. Electron-Probe X-Ray Microanalyses* of the Central Region of an Ingot of Steel (C)

	C_{max}, pct			C_{min}, pct			Segregation Ratio		
Treatment	Ni	Mo	Ti	Ni	Mo	Ti	Ni	Mo	Ti
As-cast	19.8	9.7	0.94	17.3	4.7	0.13	1.14	2.1	7.2
4 hr at 2250°F	19.2	7.8	0.54	17.6	5.1	0.22	1.09	1.5	2.5
8 hr at 2250°F	19.1	7.0	0.54	17.7	5.2	0.28	1.08	1.3	1.9
24 hr at 2250°F	19.1	7.1	0.43	17.9	5.5	0.28	1.07	1.3	1.5

*Average of eight ot more analyses.

(a) (b) (c)

Fig. 7—Photomicrographs of the central region of ingot of Steel C after homogenizing for indicated times at 2250°F and aging at 900°F for 3 hr: (a) 4 hr; (b) 8 hr; (c) 24 hr. Magnification 95 times.

to hot working. Heating at the indicated temperature for 4 hr eliminated most of the austenite from the cast and aged structure, and only occasional pools of austenite were found, Fig. 7(a). Heating for longer times (8 to 24 hr) provided sufficient homogenization to preclude the formation of these austenite pools during aging, and the structure was essentially fully martensitic, Figs. 7(b) and (c).

The degrees of segregation were determined after the specimens had been aged and given a light etch in ferric chloride solution. To express the degree of homogenization in terms of a single parameter, an index of residual interdendritic segregation (δ) was used;[9] it is defined as $[C_{\max(T)} - C_{\min(T)}]/[C_{\max} - C_{\min}]$, where $C_{\max(T)}$ and $C_{\min(T)}$ are the maximum and minimum concentrations after the thermal homogenization treatment, and $C_{\max}$ and $C_{\min}$ are the maximum and minimum concentrations in the as-cast steel. Initially $\delta = 1$, and after complete homogenization has occurred $\delta = 0$.

The results of microprobe analyses are presented in Table III and the indices of residual interdendritic segregation (δ) for nickel, titanium, and molybdenum are plotted graphically as a function of time at 2250°F in Fig. 8. It is evident that the rate of homogenization decreased in the order—titanium, molybdenum, and nickel. The initial rate of homogenization was very rapid, probably because of the steep concentration gradients in the as-cast structure; for example, the $C_{\max} - C_{\min}$ values were reduced by heating for 4 hr to 1.6 pct Ni, 2.7 pct Mo, and 0.32 pct Ti, as compared with the as-cast-structure values of 2.5 pct Ni, 5.0 pct Mo, and 0.81 pct Ti. Although partial homogenization by only thermal treatment initially was rapid and sufficient to preclude formation of reverted austenite, the rate later decreased to such an extent that complete homogenization by this treatment alone appeared to be impractical.

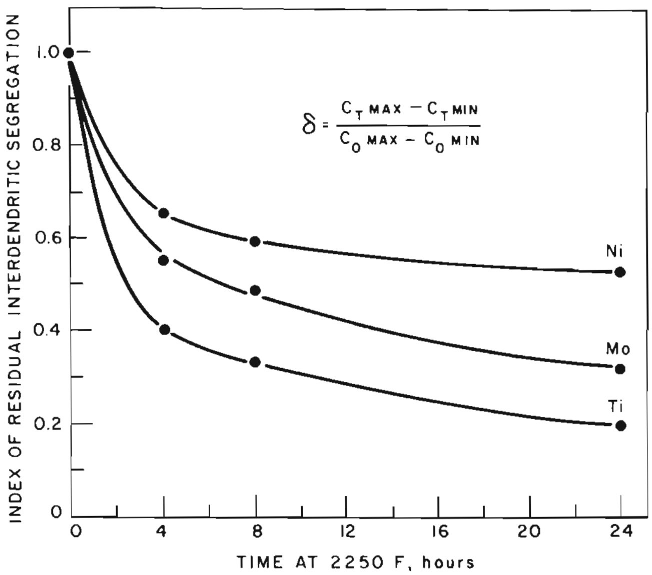


Fig. 8—Effect of time at 2250°F on interdendritic segregation in the central region of ingot C.

Table IV. Effects of Homogenization on the Mechanical Properties of As-Cast 18Ni(250) Maraging Steel (D)

Homogenization Treatment	Heat Treatment	Yield Strength (0.2 pct Offset), ksi	Tensile Strength, ksi	Elongation in 1 in. pct	Reduction of Area, pct	CVN Energy Absorbed at +80°F, ft-lb
None	1 hr at 1500°F, AC*	114	140	12.0	46.2	20.5
4 hr at 2200°F, AC	1 hr at 1500°F, AC	126	143	15.5	68.2	48
24 hr at 2200°F, AC	1 hr at 1500°F, AC	121	142	15.0	63.4	48
None	1 hr at 1500°F, AC, 3 hr at 900°F, AC	227	238	2.5	6.3	7
4 hr at 2200°F, AC	1 hr at 1500°F, AC, 3 hr at 900°F, AC	235	247	7.5	39.0	13
24 hr at 2200°F, AC	1 hr at 1500°F, AC, 3 hr at 900°F, AC	237	249	10.0	44.8	15.5

*Air-cooled.

5) EFFECT OF THERMAL HOMOGENIZATION ON THE MECHANICAL PROPERTIES OF AS-CAST AND OF HOT-ROLLED LABORATORY-CAST STEEL

A 500-lb ingot having a 7- by 11-in. cross section was teemed from an air-melted laboratory heat. The composition of the steel (D) is presented in Table I. To investigate the effect of thermal homogenization on the mechanical properties of this material in the as-cast condition, tension-test-specimen blanks (for 0.252-in.-diam specimens) and standard CVN impact-test-specimen blanks were cut from the central region of the ingot. The specimen axes were oriented parallel to the axis of the ingot. Before being tested, the blanks were homogenized at 2200°F for either 4 or 24 hr and air-cooled.

The effects of thermal homogenization on the mechanical properties of as-cast maraging steel (D) are presented in Table IV. For the steel in the annealed state, homogenizing at 2200°F for 4 hr more than doubled the notch toughness and substantially increased the tensile ductility. Heating for a longer time at 2200°F (that is, 24 hr) did not further improve the properties.

For the steel in the annealed and aged condition, thermal homogenization produced marked increases in tensile ductility and toughness (CVN) and a small but significant increase in yield strength. The increase in yield strength was probably related to the elimination of the relatively soft pools of reverted austenite that usually were present in the alloy-rich, interdendritic areas. Homogenizing at 2200°F for 4 hr eliminated most of the austenite and increased the yield strength by 8 ksi. The longer homogenization time gave a fully martensitic structure but little further increase in yield strength. Although the elimination of reverted austenite could account for the increased yield strength, it is unlikely that this change also was the reason for the marked increase in tensile ductility and toughness. Since heating to 2200°F may have altered the morphology and distribution of segregate-phase particles, an investigation of the occurrence and composition of particles present in the fracture surfaces of broken CVN specimens was conducted.

To prepare extraction fractographs from broken CVN impact-test specimens, a film of carbon was deposited on the fracture surface and then removed by electrolytically etching in a 1 pct solution of bromine in methyl alcohol. Before the homogenization, the fracture surfaces of the annealed and aged CVN impact-test specimens contained a large number of thin platelike and thin dendritic segregate-phase particles, as well as a few massive particles that were opaque to the electron beam, Fig. 9. The thin platelike and dendritic segregate-phase particles, which covered a high proportion of the fracture surface, were identified by standard electron-diffraction techniques as predominately Ti(C,N) and, to a lesser extent as Ti_2S. It has been shown previously[10] that these particles are formed by slow cooling, or by interrupted cooling, from temperatures in excess of 2000°F and that they are characteristic of embrittled maraging steel. Evidently the rate of cooling of this ingot in the mold was too slow to prevent thermal embrittlement.

After thermal homogenization, the embrittling precipitate had been almost completely eliminated, and large areas of the replica were free of particles, Fig. 10; evidently, the precipitate dissolved during the high-temperature heating, and reprecipitation of Ti(C,N) did not occur during the relatively rapid cooling of this ingot section to room temperature. Only a few massive particles, identified as predominantly Ti(C, N), were present. The lower particle density in the frac-

Fig. 9—Electron-extraction fractograph of a CVN impact-test specimen of the ingot of Steel D. Magnification 5700 times.

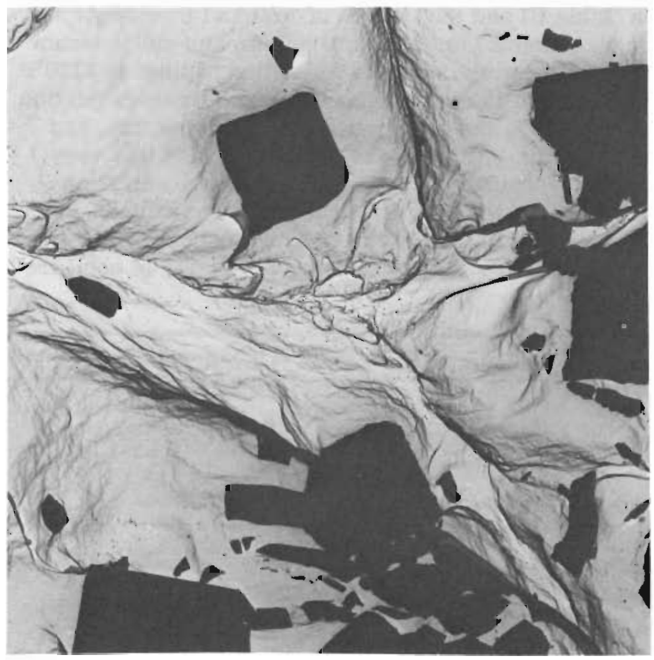


Fig. 10—Electron-extraction fractograph of a CVN impact-test specimen of ingot of Steel D that was homogenized for 4 hr at 2200°F. Magnification 5700 times.

ture surfaces of the homogenized specimens undoubtedly was a cause of the increased ductility and toughness.

To study the effect of thermal homogenization on the mechanical properties of wrought material, a portion of the 7-in.-thick ingot was straightaway-rolled to 1.9-in.-thick slab. Sections of the slab then were cross-rolled to $\frac{1}{2}$-in.-thick plate and spray-quenched, after being heated at 2200°F for either 4 or 24 hr. Standard longitudinal and transverse tension and impact specimens were cut from these plates and were tested in the annealed and in the annealed and aged conditions.

The effects of thermal homogenization on the mechanical properties of $\frac{1}{2}$-in.-thick plate are presented in Table V. No significant increases in strength, tensile ductility, or notch toughness were found as a result of the homogenization for longitudinal or transverse specimens tested either in the annealed or in the annealed-and-aged conditions. With the exception of the notch-toughness values for the material as-annealed, the properties of the longitudinal specimens were very similar to those of the transverse specimens, as would be expected for cross-rolled plate. Furthermore, the number, size, and distribution of particles in the fracture surfaces, as revealed by extraction fractographs, appeared to be independent of homogenization treatment and specimen orientation. A typical fractograph is presented in Fig. 11.

Optical micrographs showed that banding was clearly evident in the unhomogenized steel, Fig. 12, but was much less severe after thermal homogenization for 4 hr at 2200°F, Fig. 12(*b*); there was no evidence of banding in the specimens that had been heated at 2200°F for 24 hr prior to rolling, Fig. 12(*c*). In the unhomogenized steel, the difference between the composition of the bands and that of the material between the bands was 2.4 pct Ni, 3.1 pct Mo, and 0.41 pct Ti. Although this represents a severely segregated condition, the mechanical properties were very similar to those of the same steel that had been homogenized for 24 hr at 2200°F before rolling, in which banding was not evident. The differences between the maximum and minimum concentrations were less than 0.7 pct Ni, 0.6 pct Mo, and 0.1 pct Ti.

It is interesting to compare the degrees of segregation in the as-cast ingot with those in steel rolled to $\frac{1}{2}$-in.-thick plate, Table VI. The segregation in the as-cast ingot (before homogenization) was greater than that in the unhomogenized hot-rolled plate, an indica-

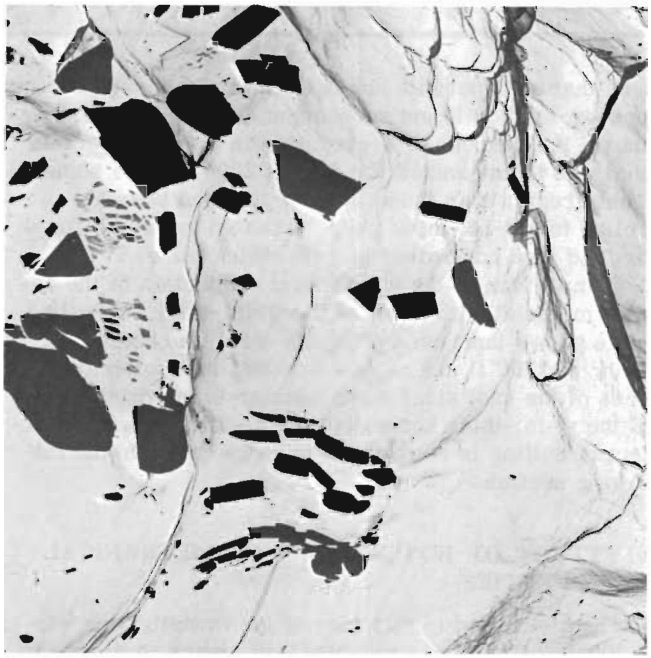


Fig. 11—Electron-extraction fractograph of a CVN impact-test specimen of $\frac{1}{2}$-in.-thick plate of Steel D that was homogenized for 4 hr at 2200°F. Magnification 5650 times.

Table V. Effects of Homogenization on the Mechanical Properties of 18Ni(250) Maraging Steel Plate*

Homogenization Treatment	Heat Treatment	Yield Strength (0.2 pct Offset), ksi	Tensile Strength, ksi	Elongation in 1 in. pct	Reduction of Area, pct	CVN Energy Absorbed at +80°F, ft-lb
		Longitudinal				
None	1 hr at 1500°F, AC†	121	140	15.0	73.4	102
4 hr at 2200°F, SQ‡	1 hr at 1500°F, AC	126	140	16.0	73.4	108
24 hr at 2200°F, SQ	1 hr at 1500°F, AC	120	138	15.0	73.8	103
None	1 hr at 1500°F, AC, 3 hr at 900°F, AC	233	246	10.0	53.4	21
4 hr at 2200°F, SQ	1 hr at 1500°F, AC, 3 hr at 900°F, AC	237	245	10.0	54.2	21
24 hr at 2200°F, SQ	1 hr at 1500°F, AC, 3 hr at 900°F, AC	238	246	10.0	55.8	21
		Transverse				
None	1 hr at 1500°F, AC	126	140	16.0	71.6	87
4 hr at 2200°F, SQ	1 hr at 1500°F, AC	126	139	16.0	73.1	85
24 hr at 2200°F, SQ	1 hr at 1500°F, AC	126	139	16.0	72.4	85
None	1 hr at 1500°F, AC, 3 hr at 900°F, AC	236	247	10.0	51.4	22
4 hr at 2200°F, SQ	1 hr at 1500°F, AC, 3 hr at 900°F, AC	239	247	10.0	54.0	20
24 hr at 2200°F, SQ	1 hr at 1500°F, AC, 3 hr at 900°F, AC	238	247	10.0	52.8	21

*$\frac{1}{2}$-in.-thick plate of Steel D.
†Air-cooled.
‡Spray-quenched.

Table VI. Effect of Homogenization and Rolling on the Segregation in a 500-lb Laboratory-Cast Ingot (Steel D)

Treatment	C_{max}, pct			C_{min}, pct			Segregation Ratio		
	Ni	Mo	Ti	Ni	Mo	Ti	Ni	Mo	Ti
No homogenization, no hot rolling	20.3	8.8	1.1	17.2	4.0	0.19	1.18	2.2	5.8
4 hr at 2200°F, no hot rolling	18.9	6.8	0.56	17.2	4.3	0.23	1.10	1.6	2.4
No homogenization, hot-rolled to $\frac{1}{2}$-in.-thick plate	19.8	7.3	0.61	17.4	4.2	0.20	1.14	1.7	3.1
4 hr at 2200°F, hot-rolled to $\frac{1}{2}$-in.-thick plate	18.8	6.3	0.47	17.5	4.7	0.29	1.08	1.3	1.6

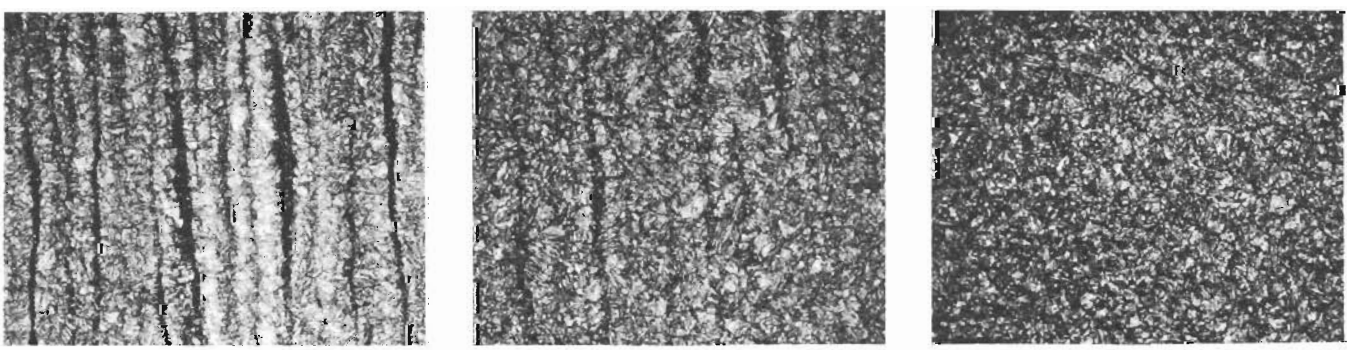
tion that considerable homogenization occurred during heating to 2200°F and subsequent hot rolling. Furthermore, the degrees of segregation in the ingot section that was homogenized for 4 hr at 2200°F were somewhat greater than those in plate that had been hot-rolled to 1.9-in.-thick plate, reheated to 2200°F for 4 hr, and then hot-rolled to $\frac{1}{2}$-in.-thick plate.

Comparison of the mechanical properties of the as-cast material with those of the $\frac{1}{2}$-in.-thick hot-rolled plate shows that, even after the steel was heated for 24 hr at 2200°F, the tensile ductility and notch toughness of the cast steel were inferior to the properties of the $\frac{1}{2}$-in.-thick hot-rolled plate. This beneficial effect of rolling is considered in more detail in the following section.

6) EFFECT OF HOT ROLLING ON MECHANICAL PROPERTIES

The steel used in this part of the investigation was a 500-lb ingot having a 7-by 11-in. cross section, and was similar to that described in the previous section. The chemical composition of this steel (E) is given in Table I. Sections of this 7-in.-thick ingot were homogenized at 2200°F for 4 hr, straightaway-rolled to plate with thicknesses of 5.9, 4.6, and 2.6 in., respectively, and then cross-rolled to thicknesses of 5, 3, and 1 in., respectively. Standard longitudinal tension-test specimens and CVN impact-test specimens were cut from central areas of the plates, and were tested in the annealed and in the annealed and aged conditions.

The results of the rolling experiments on the mechanical properties of longitudinal and transverse specimens of Steels D and E are presented in Table VII. All ingot sections were homogenized at 2200°F for 4 hr prior to rolling, with the exception of the section rolled to $\frac{1}{2}$-in.-thick plate, which was homogenized before it was cross-rolled. The reduction of area and the notch toughness of the annealed and the annealed and aged longitudinal and transverse specimens usually were greater as the degree of hot reduction was increased; however, the strength and elongation were not significantly affected. The greatest increase observed was in the notch-toughness values of the material as-annealed; the CVN energy absorption for the 1-in.-thick plate was almost twice that of the 5-in.-thick plate. This improvement in reduction of area and toughness may, in part, be related to segregation (that is, banding) as the degree of segregation in each of the plates decreased with increase in the amount of hot reduction. However, other factors must be considered. Prior-austenite-grain-size measurements on annealed plate showed that the grain size of the 1-in.-thick plate was ASTM No. 8, whereas that of the 5-in.-thick plate was ASTM No. 4. This decrease of grain size with increase in the amount of hot reduction could also have contributed to the increased toughness. Furthermore, extraction fractographs prepared from CVN specimens cut from each plate of different thickness showed a trend. As stated previously, the particles in the fracture surface of the as-cast material were not evenly distributed, but existed in well-defined groups. The effect of rolling was to reduce the number and size of the particles in these groups, with the result that, as the degree of hot reduction was increased, the particles in the fracture surface became fewer, smaller, and more evenly distributed. It is thought that the breakup of these



(a) (b) (c)

Fig. 12—Photomicrographs of sections of $\frac{1}{2}$-in.-thick plate of Steel D that had been thermally homogenized at 2200°F for indicated times: (a) 0 hr; (b) 4 hr; (c) 24 hr. Magnification 120 times.

Table VII. Effects of Hot Rolling on the Mechanical Properties of Cast 18Ni(250) Maraging Steel (D and E)

Plate Thickness, in.	Heat Treatment	Yield Strength (0.2 pct Offset, ksi	Tensile Strength, ksi	Elongation in 1 in. pct	Reduction of Area, pct	CVN Energy Absorbed at +80°F, ft-lb
		Longitudinal				
7 (as-cast)*	1 hr at 1500°F, AC	126	143	15.5	68.2	48
5	1 hr at 1500°F, AC	112	142	15.5	69.1	46
3	1 hr at 1500°F, AC	120	140	16.0	72.3	78
1	1 hr at 1500°F, AC	118	139	16.5	71.5	83
$\frac{1}{2}$*	1 hr at 1500°F, AC	126	140	16.0	73.4	108
7 (as-cast)*	1 hr at 1500°F, AC, 3 hr at 900°F, AC	235	247	7.5	39.0	13
5	1 hr at 1500°F, AC, 3 hr at 900°F, AC	237	247	10.0	49.9	13
3	1 hr at 1500°F, AC, 3 hr at 900°F, AC	239	245	10.0	50.2	19
1	1 hr at 1500°F, AC, 3 hr at 900°F, AC	236	245	10.0	51.1	19
$\frac{1}{2}$*	1 hr at 1500°F, AC, 3 hr at 900°F, AC	237	245	10.0	54.2	21
		Transverse				
5	1 hr at 1500°F, AC	113	142	15.5	66.6	44
3	1 hr at 1500°F, AC	119	142	16.0	70.8	61
1	1 hr at 1500°F, AC	117	140	17.0	72.8	82
$\frac{1}{2}$*	1 hr at 1500°F, AC	126	139	16.0	73.1	85
5	1 hr at 1500°F, AC, 3 hr at 900°F, AC	235	248	10.0	48.1	15
3	1 hr at 1500°F, AC, 3 hr at 900°F, AC	231	244	9.5	46.9	17
1	1 hr at 1500°F, AC, 3 hr at 900°F, AC	235	243	10.0	52.4	19
$\frac{1}{2}$*	1 hr at 1500°F, AC, 3 hr at 900°F, AC	239	247	10.0	54.0	20

*Steel D.

particles during rolling may also be a cause of the increased toughness.

CONCLUSIONS

1) Banding in 18Ni(250) maraging steel is a consequence of the microsegregation of nickel, molybdenum, and titanium, which originates as interdendritic segregation during ingot solidification.

2) Banding results in directional differences in plane-strain fracture toughness; the highest fracture toughness was obtained when the crack plane and direction were oriented normal to the plane of banding, and the lowest occurred when both the crack plane and the crack direction were parallel to the plane of bands.

3) Results of heating sections of small laboratory-cast ingots at 2250°F showed that the initial rate of homogenization is rapid, and may be sufficient to preclude the formation of reverted austenite during aging at 900°F. The rate later decreases to such an extent that complete homogenization by thermal treatment alone becomes impractical.

4) Heating at 2200°F had a beneficial effect on the strength, ductility, and toughness of as-cast maraging steel.

5) Hot rolling of a 7-in.-thick ingot section to $\frac{1}{2}$-in.-thick plate effected a reduction of microsegregation similar to that obtained by heating an ingot section at 2200°F for 4 hr, but resulted in even greater increases in ductility and toughness.

6) Heating 1.9-in.-thick plates at 2200°F for either 4 or 24 hr, prior to cross rolling, did not further improve either the longitudinal or transverse ductility and toughness, although the segregation was further reduced.

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