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PART II EFFECTS OF BANDING

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Thanks are also due to Messrs. R. L. Jones, J. F. Hildebrand, and F. C. Nordquist for their aid in proofreading and helping to arrange the information. In reviewing voluminous amounts of information, it is quite easy to begin to assume that principles reviewed many times are well known. Their pointing out areas that required clarification was very helpful in assuring the readability of this report to the average metallurgist.

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LITERATURE SURVEY ON BANDING IN STEEL

INTRODUCTION

During the course of investigations of several tests and service failures in components of the B-58 landing gear, it was noted that the structure of the steel was banded. Banding manifests itself as groups of light and dark etching grains adjacent to each other and strung out in the direction of rolling or forging flow, in a polished and etched metallographic specimen. These groups or rows of similar grain have the appearance of bands; hence the name banding. The presence of banding in a failed part was investigated as a cause of failure, but usually it was not severe, and, in most cases, the direction of loading was parallel to the direction of banding. Limited literature searches, made at these times, produced very few papers on the subject. In the only paper that contained strength data, the only effect of banding was a reduction of elongation properties of normalized SAE 4340 steel in the transverse direction.(17) The yield and ultimate strengths were comparable. Microhardness traverses made across the banded area of failed materials showed insignificant differences in hardness and strengths where tensile tests were conducted. In all cases it was impossible to attribute any portion of the cause of failure to banding. Some element of doubt did exist, however, and the need for knowledge of the effect of banding in steel became apparent.

In discussions with vendors and other agencies, it appeared that banding was much like the weather. Everyone knew it existed, but no one did anything about it. When the landing gear materials evaluation testing was undertaken, it was decided to conduct a comprehensive literature survey on banding, concurrent with the one on stress corrosion.

<u>Sources of Literature</u>

The <u>ASM Review of Metal Literature</u> was the most lucrative source of information on banding. Bibliographies from initially reviewed articles provided additional references. The <u>Chemical Abstracts</u>, <u>Physical Abstracts</u>, and <u>Industrial Arts Index</u> also provided many references. All were reviewed for the past five years under each listing that could be remotely connected with the subject. The bibliographies of these articles indicated that this provided a good cross section of the literature back to 1940. In addition to work done in this country, it was also found that work had been done in England, France, Switzerland, and Russia.

Outline of the Presentation

It is generally agreed that banding is the result of heterogeneities that occurred during solidification in casting the original ingot of steel. The ingot is stripped from the mold, then transferred to "soaking pits" where it is held for a period of time at 2000 - 2300 F to permit diffusion of the alloying elements to lean areas, or homogenization to occur. The ingot is then cropped (cut off) to remove areas where shrinkage has produced voids and the ingot is hot bloomed into a forging or rolling slab. In the subsequent forging or rolling operation any remaining heterogeneity is elongated in the direction of working and results in banding. During heat treatment, any difference in alloy content causes a difference in response and a resulting difference in structure. This logically arranges the discussion of banding into: (1) Solidification of the ingot, (2) Function of the alloying elements, and (3) Banded structures and their effect on heat treatment.

SOLIDIFICATION OF THE INGOT

Form and Wallace(1) discussed the solidification of metals starting with the freezing of a pure metal.

A superheated pure metal allowed to cool slowly in the container in which it was melted will gradually and evenly lose temperature until the freezing (solidification) point is reached. At this point, ideally, the latent heat of fusion will hold the temperature constant as freezing takes place. The temperature will drop after freezing with the slope of the time-temperature curve remaining constant if no phase change is encountered. The process of freezing is one of nucleation and growth. The atoms within a liquid near the freezing point tend to cluster together or form nuclei. The change in free energy ΔF is represented by:

$$\Delta F = a^3 \cdot \Delta F_v + 6a^2 \cdot \Delta F_{sl}$$

where a is the edge length of the cubical embryo, ΔF_v is the free energy change evolved from passing from a liquid to solid volume, and ΔF_{sl} is the interfacial free energy between the two phases. As shown by upper diagram in Figure 1, these opposing functions result in an increase of free energy to a certain level beyond which it decreases. The decrease in free energy beyond this point causes the reaction to occur and growth

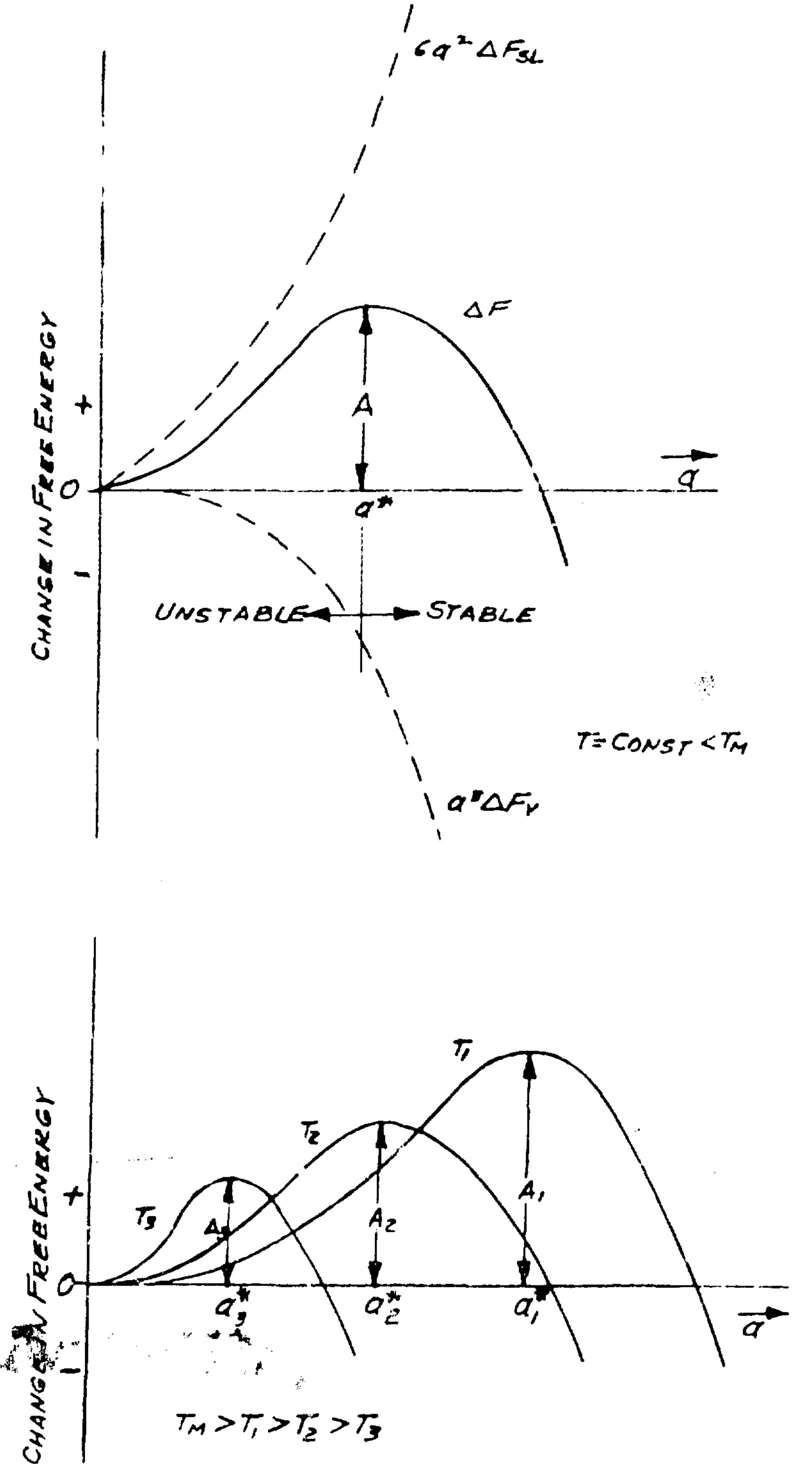


FIG 1 SOLIDIFICATION ENERGY FUNCTIONS.[1]

of the grain results. This actually causes supercooling of the liquid to a temperature below that at which it melted. As shown in the lower diagram of Figure 1, the amount of supercooling makes it possible for smaller stable nuclei to exist. This would result in a finer grain size.

The conditions of homogeneous nucleation of a pure metal are expanded to conditions of heterogenous nucleation of a pure metal as occurs in casting. When the superheated metal is poured into a cool mold, the temperature of the liquid adjacent to the wall reaches that of nucleation quickly and the fine grained chill layer forms quickly. The grains gradually become larger as the walls of the mold are heated to equilibrium with the melt and the solid-liquid interface approaches the melting point. The heat being removed from the ingot, through the mold walls and the solid layer,does not produce sufficient supercooling to permit further nucleation at this point so the already solidified grains grow inward. This results in a columnar zone in the ingot. Depending on mold configuration and temperature gradient, the temperature of the liquid will again be supercooled to a point where nucleation can occur and the remaining liquid will again nucleate and grow equiaxed grains. The proportions of chilled, columnar, and equiaxed zones will vary from metal to metal.

The solidification of alloys causes the situation to become complicated because the melting point does not remain constant. In the case of a solid solution freezing from a homogeneous melt of composition A, the freezing temperature of the initial nuclei would be "T_o" as shown in Figure 2.(3) The composition of the nuclei is B. As the nuclei grows and the temperature is lowered the composition of the melt becomes enriched and the composition at the solid-liquid interface becomes enriched. The last of the melt solidifies at temperature T_f and the composition of this portion is C. If the melt solidified evenly in a single crystal, the composition at the center would be B and that of the outer surface would be C. This is known as coring. In casting this alloy in a mold, the chill layer would freeze too rapidly to show any of this effect. As the solidification rate decreased, there would be sufficient time for this coring to occur.

The structure of the columnar zone of an alloy is not perfectly columnar. The grains usually develop as dendrites which can be described as crystals having tree-like patterns composed of many branches projecting in all directions. These dendrites are lean in alloy content and the areas between them are rich. In addition, the concentration in the center

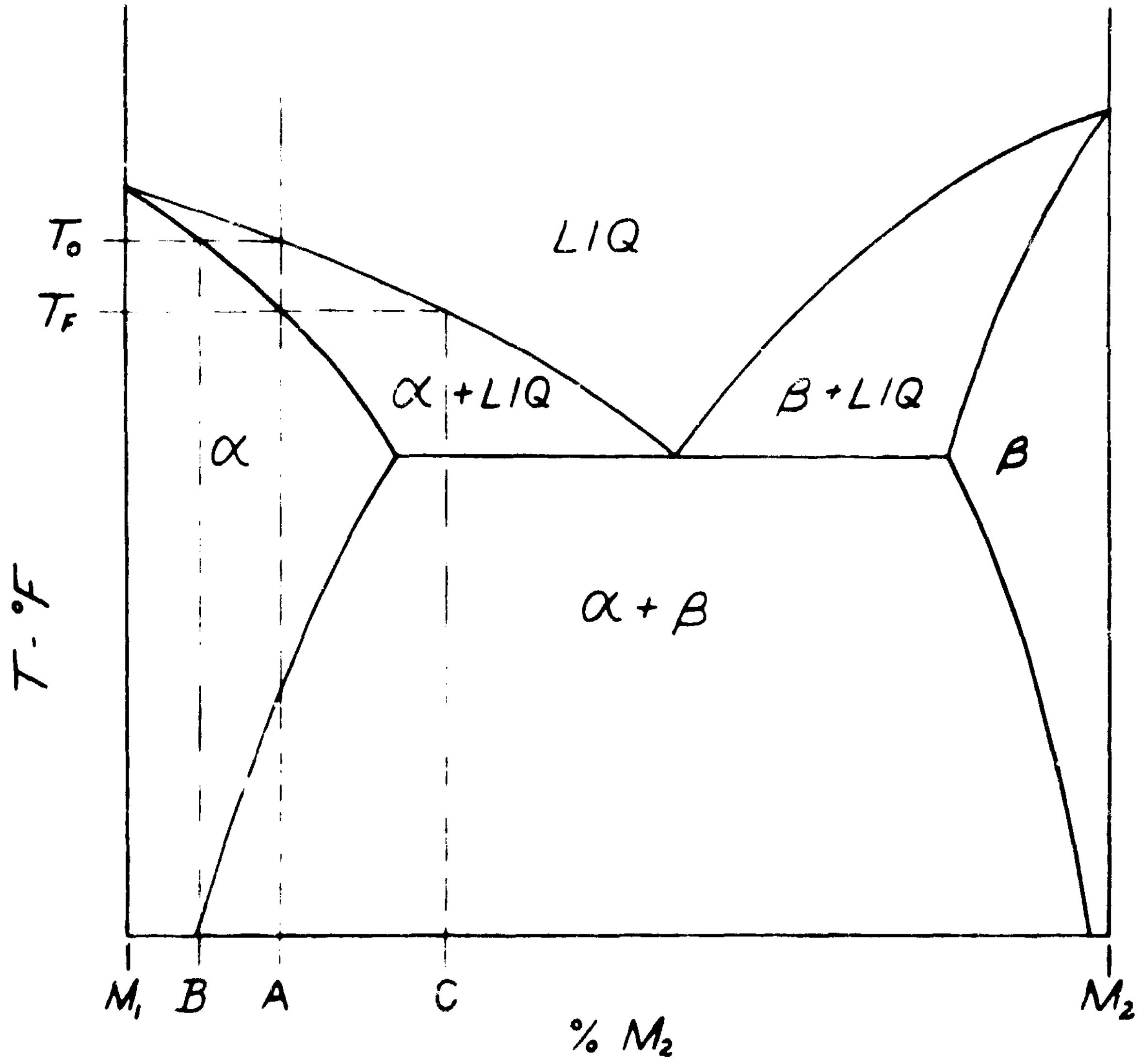
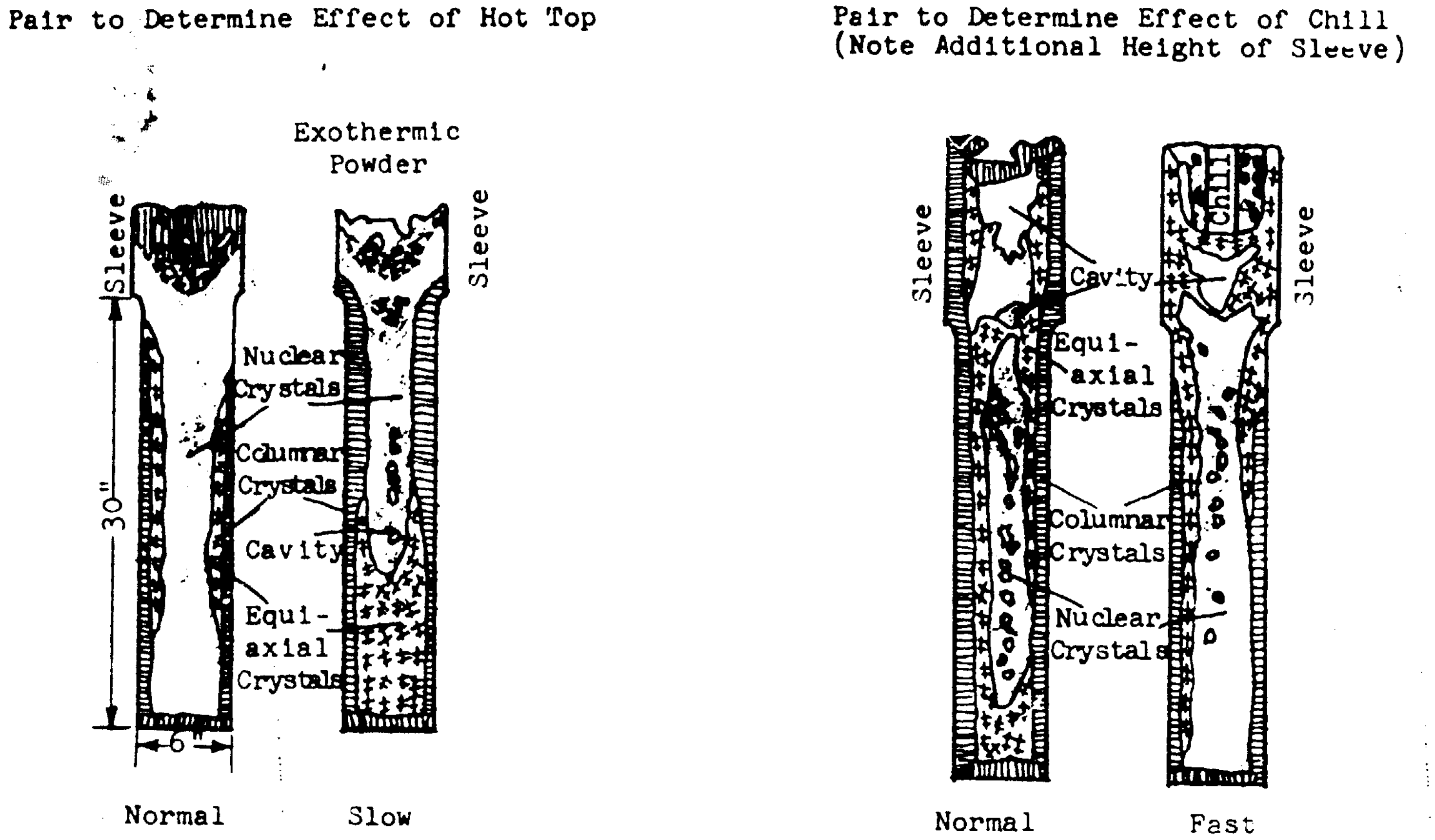


FIGURE 2. Typical Phase Diagram For Two Metals

of the ingot is richer than the surfaces. The enrichment of the interdendritic areas is the result of coring during growth of the dendrites and entrapment of the liquid that becomes enriched as the coring develops. The natural solute enrichment of remaining liquid causes the last area to freeze to be more concentrated in that element. The interdendritic segregation is termed microsegregation while the surface to center segregation is macrosegregation.

Another cause of macrosegregation is gravity or the "falling crystallite" (2, 4) theory. The change in volume on freezing is a contraction for most metals. This increases the density of the solidified nuclear crystals and they tend to sink to the bottom of the mold. The natural rise of heat and heat loss through the bottom of the mold add to this effect producing a top to bottom segregation in addition to the surface to center.

One paper(2) dealt with cooling rate and its effect on the resulting structure of a .35 carbon steel. The cooling rate of six inch diameter-30 inch long ingots was varied using exothermic powder and chills in the sleeve or riser at the top of the ingot to decrease and increase cooling rate. As shown in Figure 3, columnar zones of the slabbed ingots were increased by slower cooling in each pair of ingots poured from a common



Pair to Determine Effect of Hot Top

Pair to Determine Effect of Chill
(Note Additional Height of Sleeve)

FIGURE 3 - DIAGRAMS OF MACROSTRUCTURE IN VARIABLY COOLED CASTINGS (2)

header. The equiaxed zone, or zone of larger equiaxed crystals, was found at the bottom of the mold and increased in size with slower cooling. The columnar zone of the slowest cooled ingot was also thinner at the expense of equiaxed crystals indicating fallen crystallites formed this region while the columnar growth was taking place in the upper portion of the ingot. The remaining portions, nuclear crystals, were equiaxed, but not large grained, indicating that they all grew and solidified under similar conditions, probably at the same time. The degree and area of segregation were unmistakable.

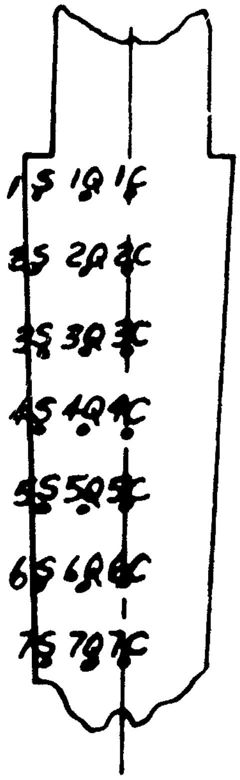
Although porosity and cavities do not directly affect banding, they are a very real and closely related problem. Note that the casting in Figure 3 that had the most homogeneous structure also had the least porosity. This ingot could be cropped at the sleeve and the remainder would be a porosity free ingot ready for further processing. In this case the mold design and pouring conditions produced a sound ingot which had minimum segregation. There are many variables that enter into the design of an ingot mold; some help to minimize segregation while others promote it. In order to produce a porosity free casting, also free of a pipe or shrinkage cavity downward from the top of the ingot, hot tops are used similar to the sleeve shown in Figure 3. The hot top tends to slow down cooling and promote segregation. To counteract slow cooling, salients or involutes are sometimes built into a round mold to increase the cooling rate of the ingot.

QUANTATIVE DATA ON SEGREGATION

In work reported at the regional meeting on the quality requirements of Super Duty Steels(4), one paper was presented on segregation and ingot structure. A summary of this paper follows:

The ingots used in the test were a 16 inch square - 3,200 pounds, 19 inch round - 3,950 pounds, 20 inch square - 4,390 pounds, 21 inch round - 6,160 pounds, and 24 inch square - 9,840 pounds. Fully killed SAE 4340 steel ingots of each type were poured at the normal casting temperature. An additional 21 inch round ingot was poured from the same heat as the others, but after pouring the melt into a holding ladle to lower the temperature 100 F. The ingots were cut an inch from the center line, machined flat, and macroetched. The ingots poured at the normal temperature all had distinct chill zones adjacent to the mold wall; a zone of columnar grains grown perpendicular to the mold wall; and a central zone of randomly oriented small dendrites, which were finer towards the center and bottom of the ingot. There was some "herring bone structure" toward the top center portion of the ingot. These are hot tears that occur in the last portion to freeze and their length was related to the ingot length-diameter ratio. These are of little consequence when sufficient hot work was done on the ingot to weld them, or "heal" the porosity. Drillings were taken for chemical analyses of the alloy and "tramp" elements at seven equal heights at the surface, center, and quarter thickness of the ingot. A summary of the results is shown in Table I. The presence of segregation was determined by calculating an allowable error according to ASTM standards. The minimum and maximum allowables were determined by adding and subtracting the error from the average of all tests made on each ingot. Contents greater than the maximum were considered a positive segregation of the element and less than the minimum, negative. There was some positive carbon segregation indicated on the 16 inch square ingot, but this was barely above the maximum (.430>.429) and adjacent analyses did not appear to indicate any trend. The evidence of phosphorous and sulfur segregation in the 20 inch square ingot was also very small and can be considered academic since the contents were small (in the range of .020% compared to .040% maximum for aircraft quality SAE 4340). The most significant

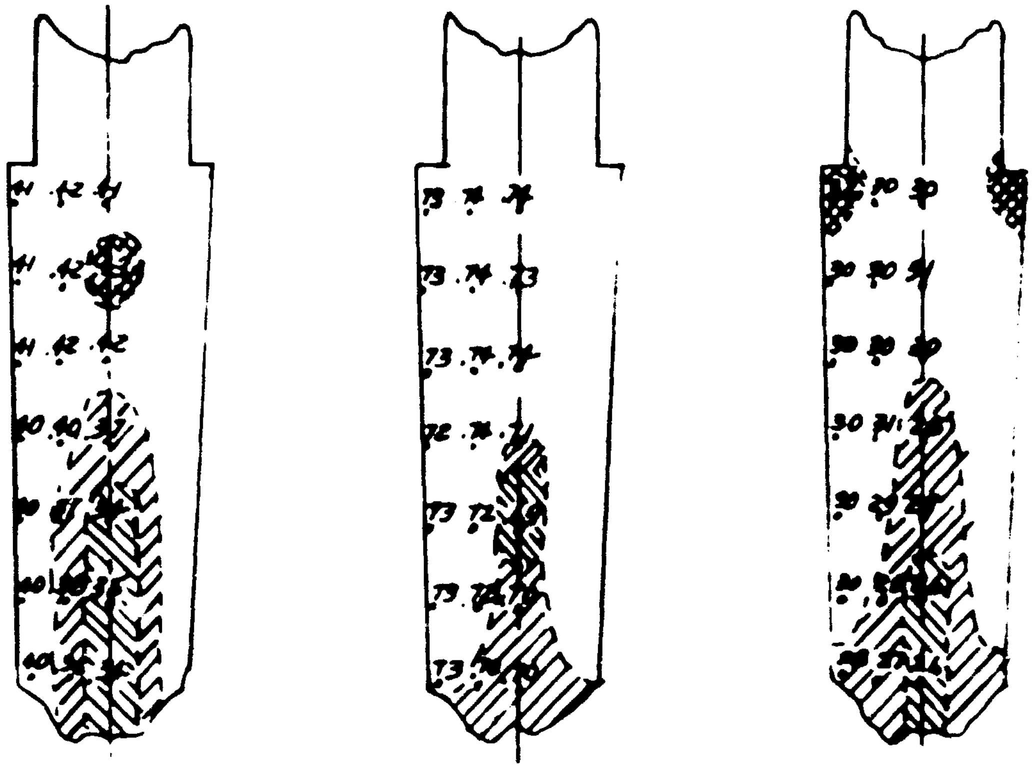
TABLE 1. SUMMARY OF CHEMICAL SEGREGATION ENCOUNTERED IN INGOTS(4)



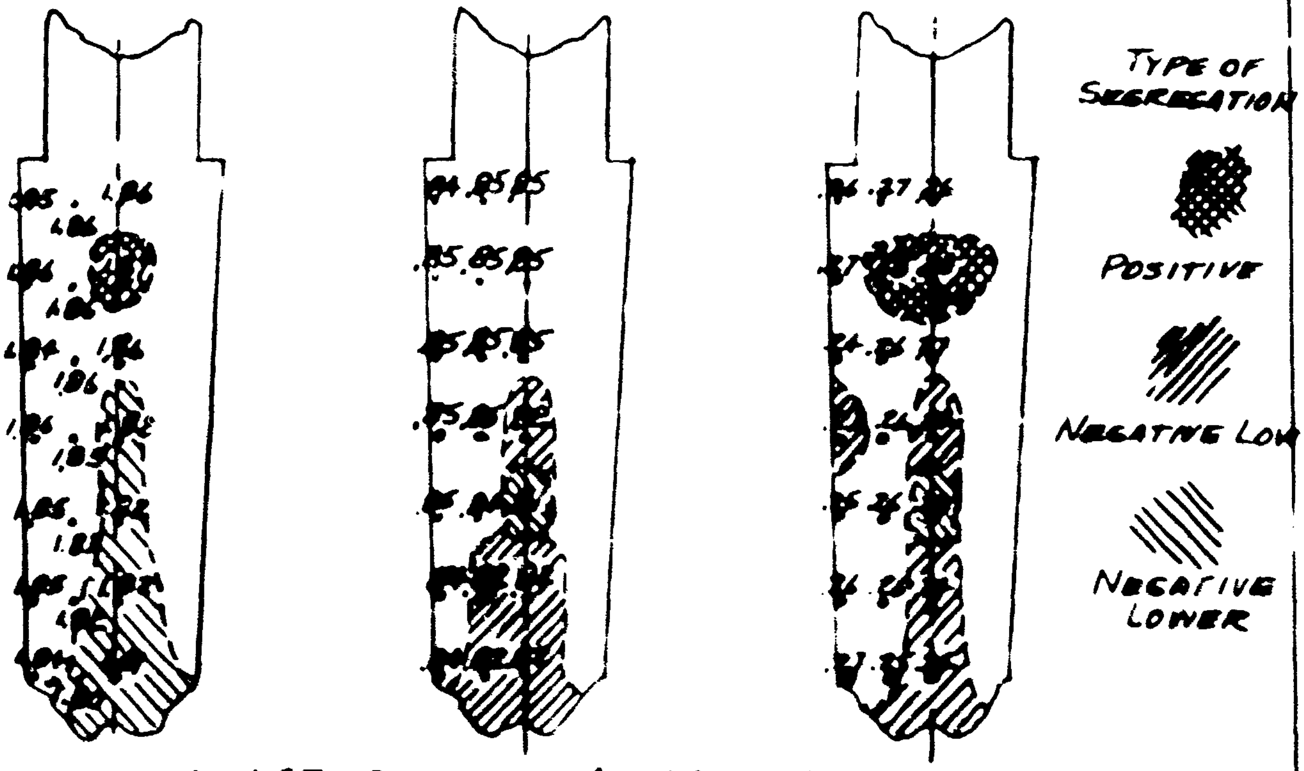
Ingot Size & Segregation	Element							
	C	Mn	P	S	Si	Cr	Ni	Mo
16 In. Square-3200 Lb.								
Positive	4Q							
Negative								
19 In. Round-3950 Lb.								
Positive								
Negative								
20 In. Square-4930 Lb.								
Positive			7C	3C				
Negative			2Q					
21 In. Round-6160 Lb. Normal Pouring Temp								
Positive								
Negative								
Low Pouring Temp								
Positive	2C				1S			2Q,2C
Negative	5,6,7Q 5,6,7C	2Q 5,6,7C			2S,6,7Q 4,5,6,7C	6,7Q 4, ,6,7C	7Q;4, 5,6,7C	

segregation occurred in the 21 inch round ingot which was cast cold. A detailed analysis of this ingot is shown in Figure 4. Note that there is negative segregation in the middle bottom for each element. This was probably due to "falling crystallite" phenomenon caused by the colder liquid tending to form both more and larger stable nuclei as explained previously. This enriched the remaining liquid, and the carbon, nickel, and molybdenum segregated positively in the last area of the ingot to freeze. This area had a herring bone appearance in the macrostructure. Hot tearing relieved the shrinkage stress formed during solidification, and directional thermal contraction caused the herring bone pattern.

A further investigation was made to determine the effect of pouring temperature on carbon segregation in medium carbon, low alloy, open hearth steels. These were 21 inch round - 6,760 pound ingots which were cropped, top and bottom, and then hot sheared into forging blooms. The top and bottom blooms were sampled and checked for carbon content in the first and last ingot from each heat. Seventeen heats were poured cold, ten heats poured at normal temperature, and seven heats poured hotter than normal. The carbon contents varied from .20 to .63% and alloys were of the 10XX, 13XX, 31XX, 40XX, 41XX, 43XX, 46XX, 50XX, 51XX and 86XX families. The heats poured hot and normal exhibited a .015 to .021% difference in carbon content from top to bottom in both the first and last ingots, while



CARBON AV. .404 MANGANESE AV. .729 SILICON AV. .300



NICKEL AV. 1.85 CHROMIUM AV. .846 MOLYBDENUM AV. .254

FIG 4 DETAILED ANALYSIS OF 21 IN. ROUND INGOT POURED AT LOW CASTING TEMPERATURE (4) the heats poured cold showed a .036 to .050% difference. This was the only trend noted. Since drillings were analyzed to determine the composition, the carbon content variation must be considered a measure of the macrosegregation.

The microsegregation of alloying elements in steel were investigated by Kohn and Philbert(5) in France using the "Castaing" microprobe. Samples were taken from a 100 ton heat of a Ni-Cr-Mo steel, similar to SAE 4335 but with slightly higher nickel and lower chromium contents. The body of the dendrites in the cast structure and the interdendritic areas were analyzed using the microprobe. The results were:

Element	Alloy Content - Wt. %		
	Mean	Dendritic Body	Interdentritic Spaces
P	0.022	----	0.090
Ni	2.5	2.4	3.25
Cr	0.66	0.55	1.05
Mo	0.29	0.13	0.95
Mn	0.50	0.50	0.20

The average concentration of the major alloying elements in the interdendritic spaces was richer than the mean content indicating positive microsegregation. Contrarily, the dentritic bodies were leaner and hence areas of negative microsegregation. Phosphorous was completely rejected from the dendritic bodies, whereas the interdendritic spaces exhibited lower manganese.

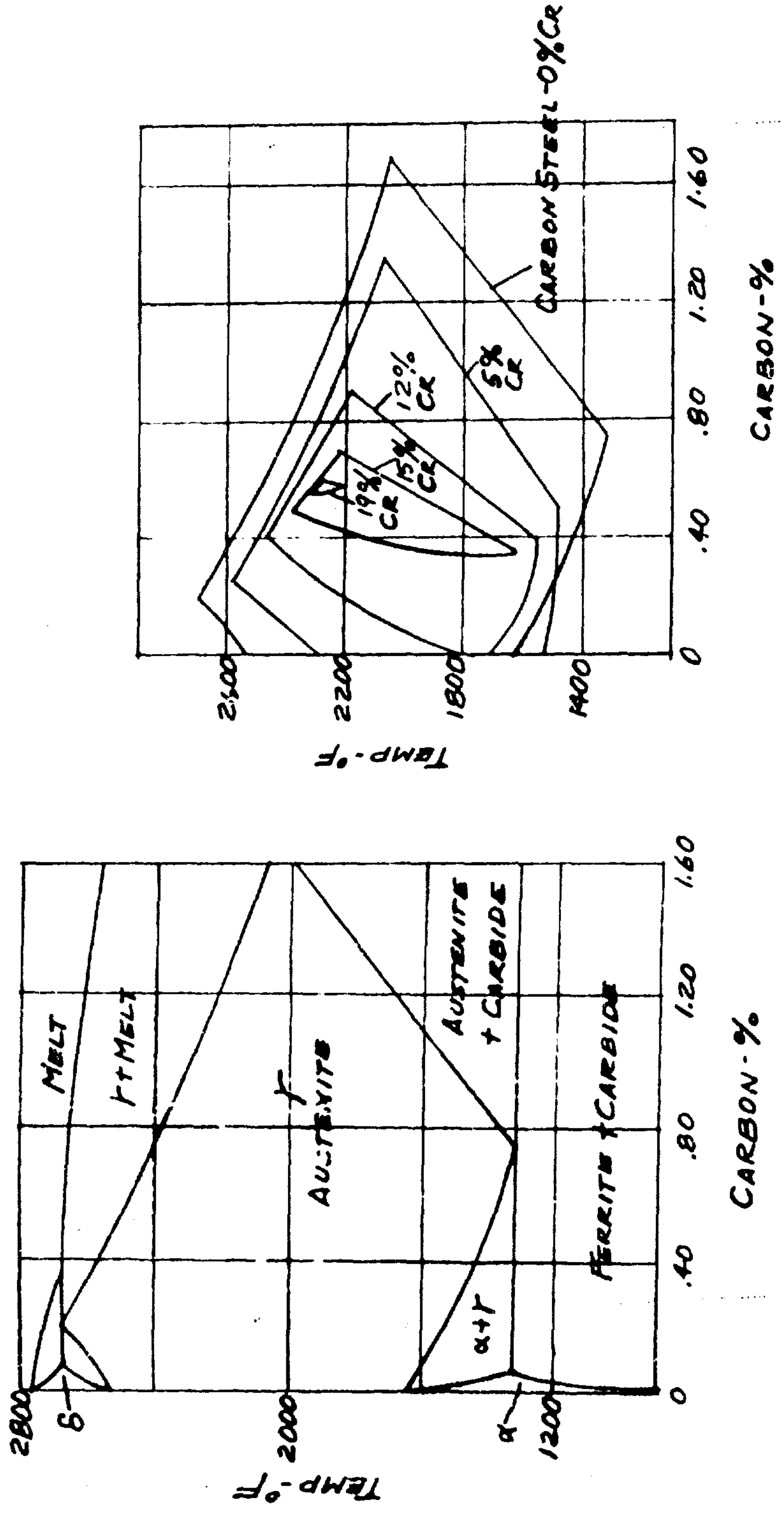
FUNCTION OF THE ALLOYING ELEMENTS

Carbon is the major alloying element in steel. The heat treatment of steel is based on the greater solubility of carbon in face-centered cubic austenite or gamma iron, and the decreased solubility in the body centered cubic ferrite or alpha iron. As shown in Figure 5a(6), the temperature at which this occurs varies with carbon content.

Changes in the shape of the iron-carbon diagram affected by the addition of elements are used to explain the characteristics of various steels. For example, chromium shrinks the austenite loop, and shifts the temperature and composition of the eutectoid as shown by Figure 5b.

The effect of various elements on a .40% carbon steel have been determined and are shown by Figure 6. Manganese and nickel depress the transition temperature, A_3, at which the transformation from ferrite to austenite is complete, while the other elements raise it.

Phosphorous and sulfur, considered tramp elements in most cases, are present in the ores. During the refining processes, they are minimized, but what does remain has a profound effect on steel. Usually controlled to .05% maximum, phosphorous(8) has a low solubility in both alpha and gamma iron and tends to segregate. While it does not form carbides, it has a great effect on the hardenability of the steel. Sulfur, generally



(b) EFFECT OF CR ON THE AUSTENITE REGION OF THE Fe-C DIAGRAM.

(a) Fe-C DIAGRAM FOR COMMERCIAL STEEL

FIG 5 · IRON-CARBON EQUILIBRIUM DIAGRAMS (6)

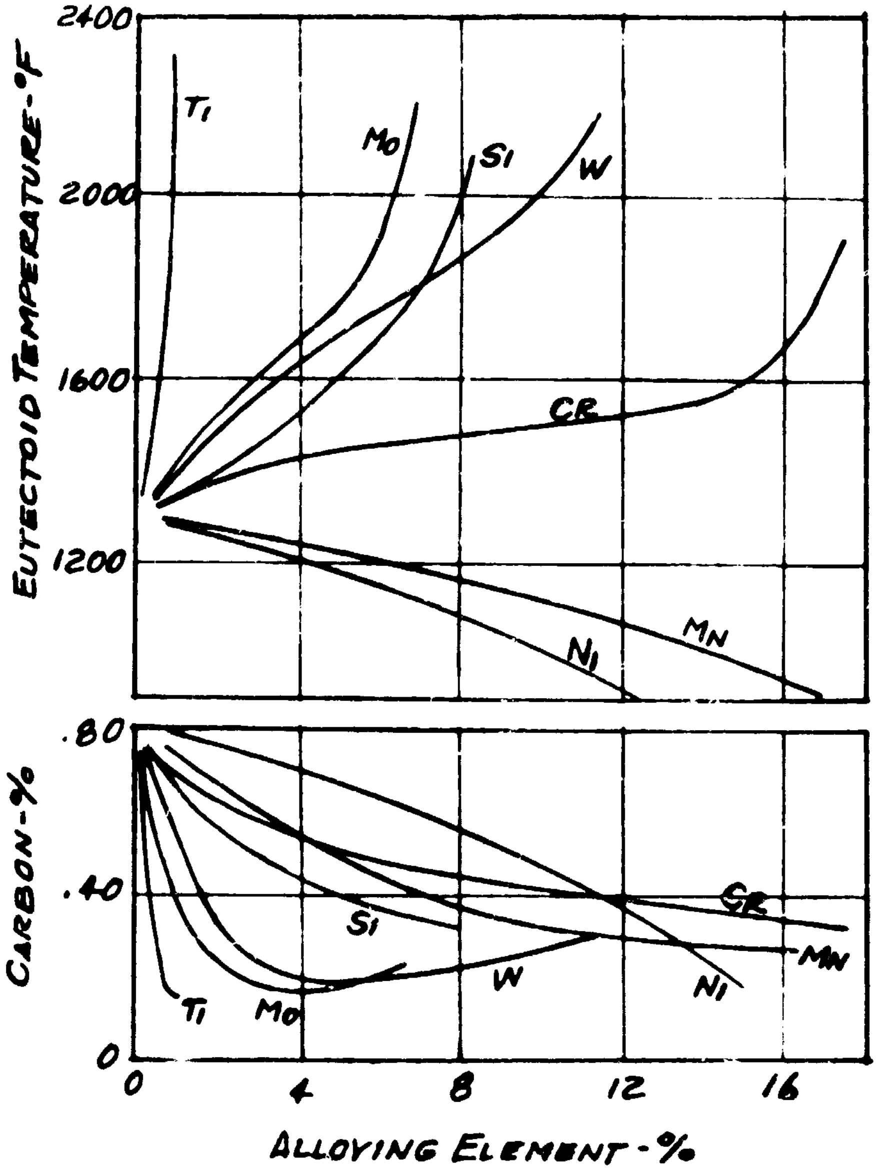


FIG 6 EFFECT OF ALLOYING ELEMENTS ON THE EUTECTOID COMPOSITION AND TEMPERATURE OF A .40C STEEL(7)

controlled to an .05% maximum, combines with iron as an intermetallic compound which has a low melting point and causes hot shortness.

The manganese present in most steels has a great affinity for sulfur, but the manganese-sulfur intermetallic compound formed does not cause hot shortness. In fact, sulfur can be added, if sufficient manganese is present, to improve the machinability without promoting hot shortness as is done in "free machining" steels. Manganese, in addition to lowering A_3, moderately increases the hardenability of steel. It strengthens ferrite and has a moderate tendency to form carbides.

Nickel which lowers the A_3 and mildly increases the hardenability, contributes greatly to the toughness of steels and lowers the transition temperature for ductile-brittle fracture. It has mild carbide forming characteristics and increases the strength of ferrite.

Chromium, molybdenum and vanadium are considered carbide formers. Chromium promotes hardenability, improves toughness, and increases the corrosion and oxidation resistance. The stability of the chromium carbides increases the tempering temperature of steels. Molybdenum increases the hardenability and also increases tempering resistance. Vanadium is the strongest carbide former of the three elements. In addition it promotes a fine grain size during solidification of castings and also increases the grain coarsening temperature of a steel during heat treatment.

Silicon has a negative carbide forming tendency. Its effect on hardenability is moderate, but it strengthens the ferrite. Silicon, along with manganese and aluminum, is an efficient total deoxidizer(9) for steel. Carbon, of course, is the prime deoxidizer. When a heat of steel is tapped from the furnace the excess carbon in the melt reacts with any oxides present. The gases evolved produce a churning action in the ladle that is desirable in mixing the ladle additions homogeneously throughout the melt. The other deoxidizers are added last and combine with the remaining oxides to eliminate the last traces. As the evolution of gas ceases, the carbon content is stabilized and a killed heat is ready for pouring.

BANDED STRUCTURES IN STEEL AND THEIR EFFECT ON HEAT TREATMENT

Older Work (Prior to 1940)

In 1936 Prohoroff surveyed the literature related to banding in ferrous alloys.(10) The first iron alloy mentioned in this connection was wrought iron. This alloy is a mixture of essentially pure iron with about 2% slag distributed throughout the matrix. In rolling, the slag inclusions are strung out in a more or less banded appearance. This gave rise to the first explanation of ferrite banding proposed by Brearly(11) in 1909. He suggested that elongated inclusions act as nuclei on which ferrite was precipitated from austenite as steel cooled through its critical range. Later in 1914 Stead(12) showed that phosphorous could cause ferrite banding in steel, presumably by forcing carbon out of interdendritic spaces into the purer dendritic area. In 1918 Le Chatelier and Bogitch(13) suggested that segregation of oxygen rather than phosphorous caused banding in steel. In 1934 Johnson and Buechling(23) reported that four hours at 2050 F was sufficient to completely remove ferrite banding, but it reappeared in tubing pierced from the bars.

These theories and observations represented the state of the art when Prohoroff's work was started. He investigated the influence of non-uniform distribution of some of the alloying elements in steels as a function of slow and fast cooling through the critical range. Pieces of steel with different compositions were heated and forged to 1/4 inch thickness. The forged strips were cut into specimens which were heated above the critical temperature, then cooled at different rates. Comparison of "as forged" with slow and fast cooled specimens showed that: (1) On heating, as revealed in the fast cooled specimen, silicon diffused into a carbon rich area and carbon diffused out, but on slow cooling the carbon migrated to areas low in silicon; (2) on slow cooling carbon migrated to areas rich in manganese or nickel; and (3) at high temperatures carbon diffused to areas rich in titanium to form stable titanium carbides which, on slow cooling, did not permit a rediffusion of carbon out of these areas.

In 1940 Sachs' (14) opinion on banding was that: ". . .Coring is a general feature in any steel (appearing as ferrite bands) but is more persistent in alloy steels than in carbon steels and particularly so in Ni-Cr steels. In steel, the various elements have different rates of diffusion and therefore homogenize differently. Phosphorous, for instance, diffuses very slowly, and is the source of the banded structure of commercial steel products."

In some work done by Williams(15) in the same year, five heats of SAE 3140 were compared on the basis of the Jominy End-Quench Hardenability. Their chemical compositions, which were similar were:

Steel Code No.	Composition - %								Gr Size
	C	Mn	Si	P	S	Ni	Cr	N	
S	.44	.84	.021	.019	.032	1.18	.67	.0053	6
T	.42	.78	---	.026	.025	1.29	.52	----	7
V	.40	.71	---	.015	.026	1.20	.63	----	7
W	.40	.74	---	.021	.028	1.19	.65	----	7
X	.40	.70	.022	.017	.025	1.32	.54	.0067	6

All of the .40 carbon steels used in this investigation were annealed at 1600 F prior to testing. The hardenability of the steels after austenitizing at 1500 F for one half hour was:

Rockwell Hardness - C Scale

Steel Code No.	Distance From Quenched End - Inches 1/16	1/2	1	1-1/2	2-1/2
S	55	49	34	32	32
T	57	45	32	29	29
V	55	41	30	26	19
W	54	40	31	28	19
X	52	30	22	17	13

The hardness for steels S and T was definitely higher than the other three steels, particularly at longest distances from the quenched ends. Note that both steels exhibited no change in hardness at 1-1/2" and 2-1/2" distances, whereas the other three steels showed a decreasing hardness at these distances. These data indicated that steels S and T had a higher hardenability.

The author also tried heating for one and three hours at 1500 F which slightly increased the hardenability of these steels. He then increased the austenitizing temperature to 1700 F for one hour. The hardenability curves for steels V and W fell almost on top of steel T with this treatment, however, steel X was still low. When steel X was held for eight hours at 1700 F, hardness of 26 and 19 was obtained at the two farthest points. A copy of the photomicrographs of identically annealed samples of each steel is shown in Figure 7. The author felt that the hardenability behavior is in line with the degree of banding and that banding is a manifestation of nonhomogeneity and resistance to diffusion. He also states that microscopic examination of quenched specimens showed a decrease in banding as maximum

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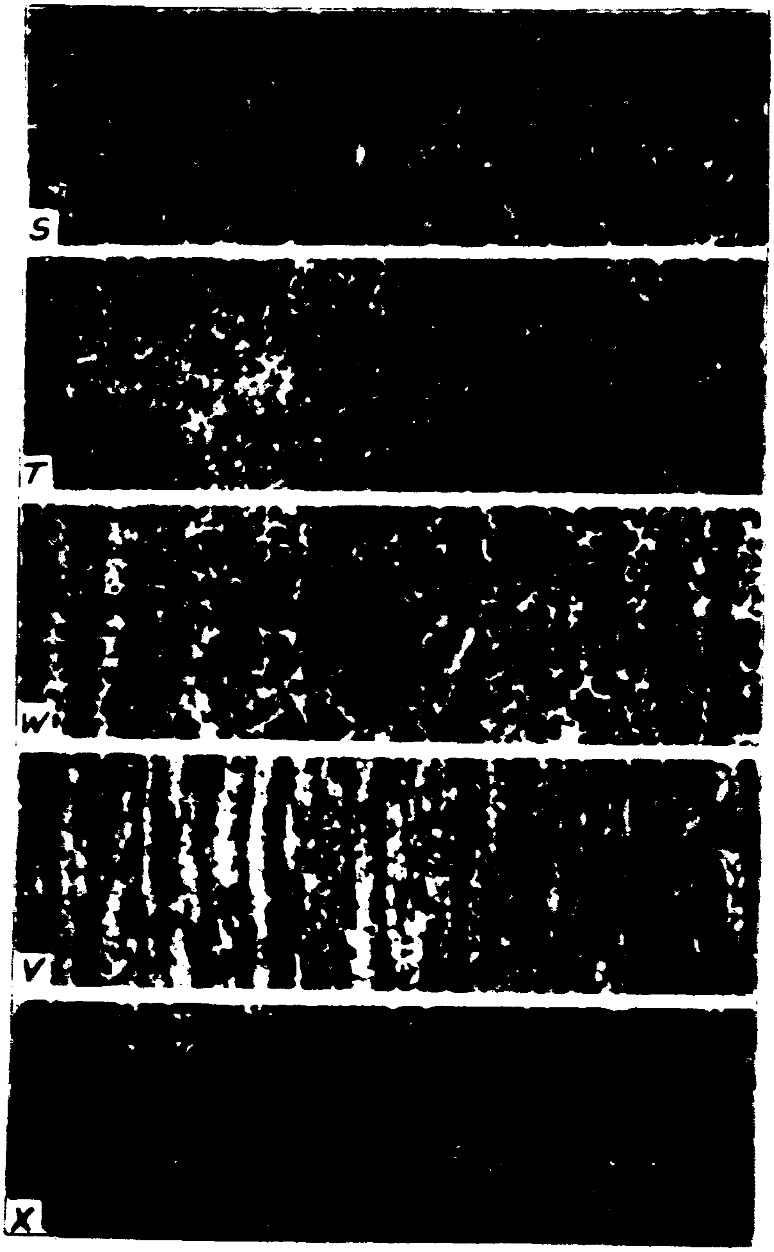


Fig. 13—S.A.E. 3140 Steels. Annealed 1 Hour at 1600 Degrees Fahr. Furnace Cooled. Time Required to Reach Maximum Hardenability Increasing Toward Bottom. × 100

FIGURE 7. Microstructure of Steels Used by Williams (15) Reduced Approximately 1/2 in Reproduction.

hardenability was approached.

In discussing the effects of temperature, the author pointed out that in one steel which showed very slow attainment of maximum hardness, 10 hours at 1500 F, 2.5 hours at 1550 F and 1/2 hour at 1600 F gave identical results. The effect of 1/4 hour at 1650 F was greater than one hour at 1600 F. This points out that 8 hours at 1700 F should be many times more effective than 1/2 hour at 1500 F, yet steel X, after this treatment, didn't attain the same hardness as steels S and T.

Note in the following table that there is little difference in the carbon and phosphorous contents of the five steels, although steels S and T are slightly above the nominal amount of carbon.

Composition and Composition Ratios - %

Steel No.	C	P	Ni/Cr	Ni+Mn	Ni+Mn / Cr	D.I.*
S	.44	.019	1.76	1.51	2.26	9.48
T	.42	.026	2.47	1.30	2.50	9.86
V	.40	.015	1.91	1.34	2.10	9.00
W	.40	.021	1.83	1.39	2.14	9.[illegible]0
X	.40	.017	2.43	1.24	2.30	9.89

*D.I. is the critical ideal diameter; that diameter bar which if quenched would have its center 50% transformed to martensite.(16)

Phosphorous has been suggested as the cause of banding, but the phosphorous content of the three banded steels (V, W, X) fell in the same range as the unbanded ones. Nickel and manganese are austenitizers and chromium is a carbide former and ferritizer. Since the effect of these elements is opposing, the ratio of the two is often used to indicate a steel's reaction to heat treatment. As can be seen from the Ni/Cr and (Ni+Mn)/Cr values, the X steel ratios were not significantly different from those of the other steels. In fact, the ratio for steel X fell between the ratios for steels S and T as did the critical ideal diameter (D.I.) index of hardenability.

Comparison of the microstructures in Figure 7 show the only significant differences in the steels that might account for their reactions to the heat treatments. Steel S, which had good hardenability, exhibited ferrite thinly and evenly dispursed around the grains. Steels W and V had large areas of ferrite which would require more time at temperature, or higher temperature, to diffuse the alloying elements within the dark grains throughout the ferritic areas. The structure of steel X, however, is more or less a mixture of the two types. The ferrite is thinner and more well dispursed than W and V, with some evidence of banding. This suggests that something was impairing the diffusion of the alloying elements. This source of impairment was on a microsegregation basis.

Present Work (Since 1940)

The most extensive work done on banding was reported in 1956.(17) Five .40 carbon steels were used in the investigation:

Type A.I.S.I.	Major Alloyed Elements	Chemistry of Heats Investigated C	Mn	Si	Cr	Ni	Mo
1340	Mn	.39	1.70	.27	.16	.29	.03
2340	Ni	.41	.68	.29	.23	3.20	--
4140	Cr-Mo	.39	.82	.31	.96	.13	.20
4340	Ni-Cr-Mo	.42	.72	.29	.82	1.74	.25
5140	Cr	.42	.85	.27	.81	----	--

Several techniques were used to bring out the banded structure of these steels, but most effective was isothermal transformations at the nose of the TTT (Time-Temperature-Transformation) curve and holding. Curves for the five steels are shown in Figures 8 - 12(18). A survey on the 4340 indicated that transformation at the upper nose of the transformation diagram brought out the banded structure most effectively. All of the steels were austenitized, quenched, isothermally held for periods of time, and finally quenched to room temperature before examining microscopically. All steels developed banded structures. Portions of the steels were homogenized at 2200 F, then normalized twice at 1700 F and three times at 1550 F to refine the grain size. The steels were austenitized and transformed at or near the nose of the TTT diagram for times to produce partial and complete transformations. The 1340, 4140, and 5140 steels, completely homogenized after 25 and 100 hours treatments, showed no evidence of banding. The 2340 steel, in which nickel is the major alloying element, showed faint traces of banding after the 100 hours homogenization. This is in general agreement with the diffusion rates for the major alloying elements:

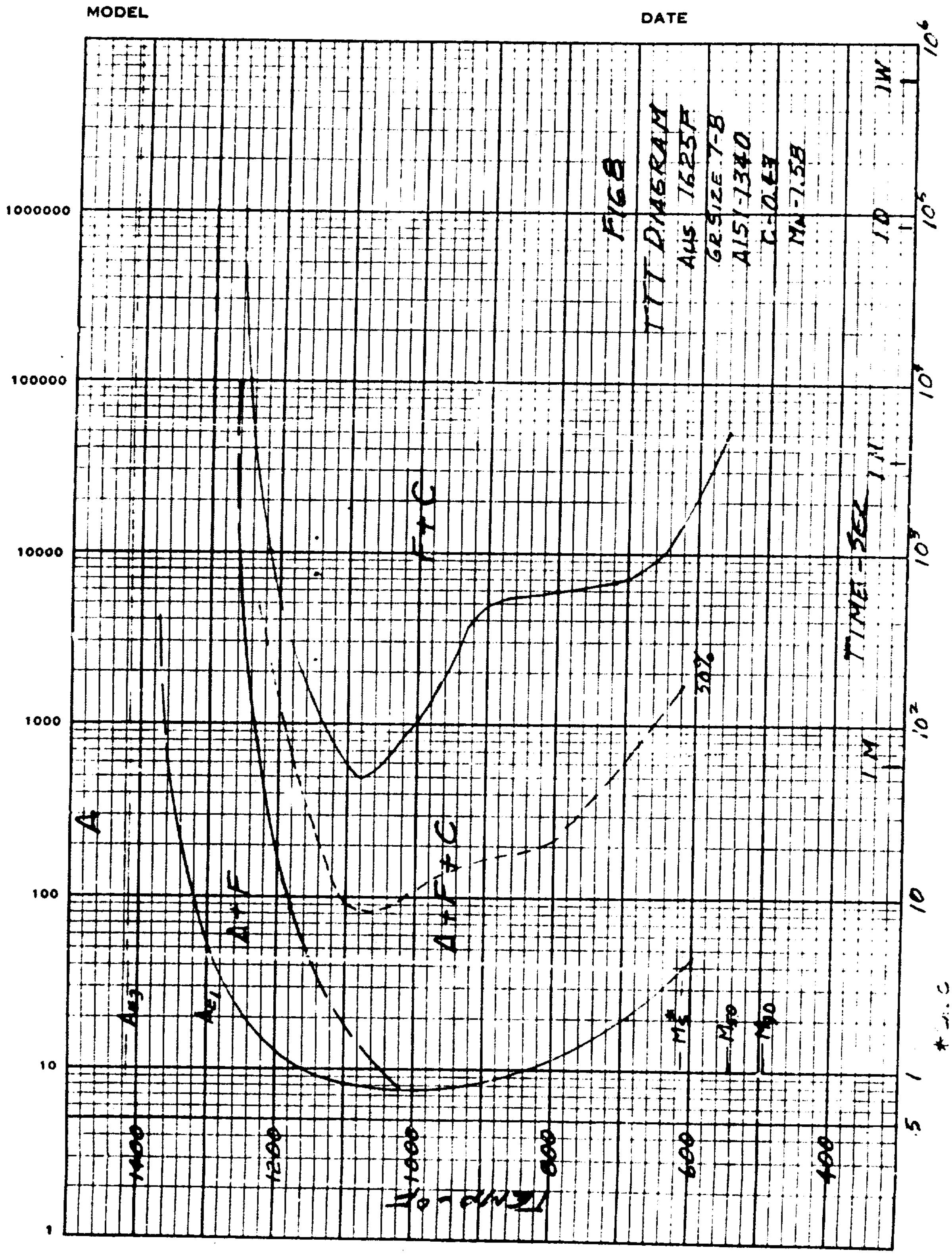
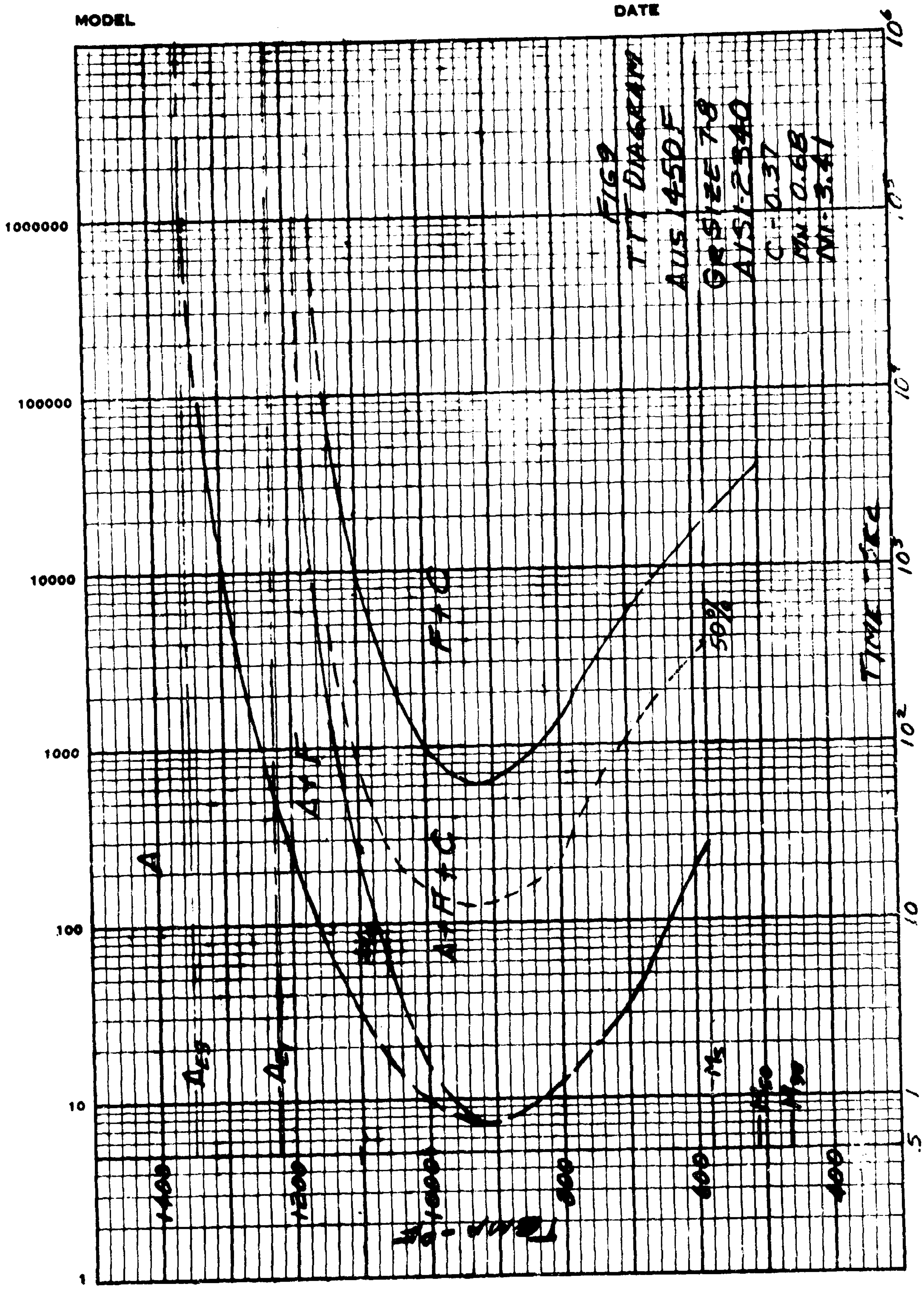
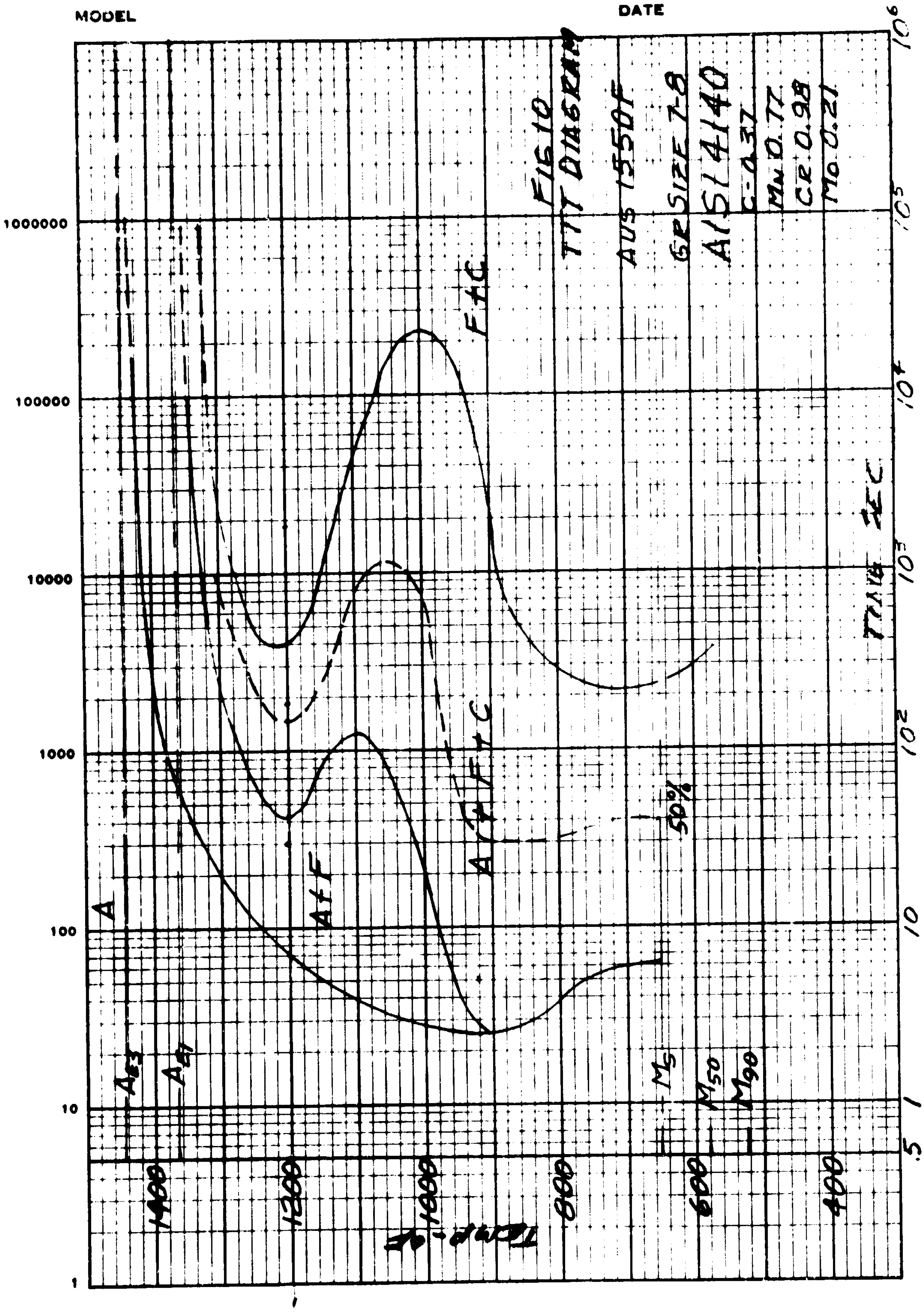
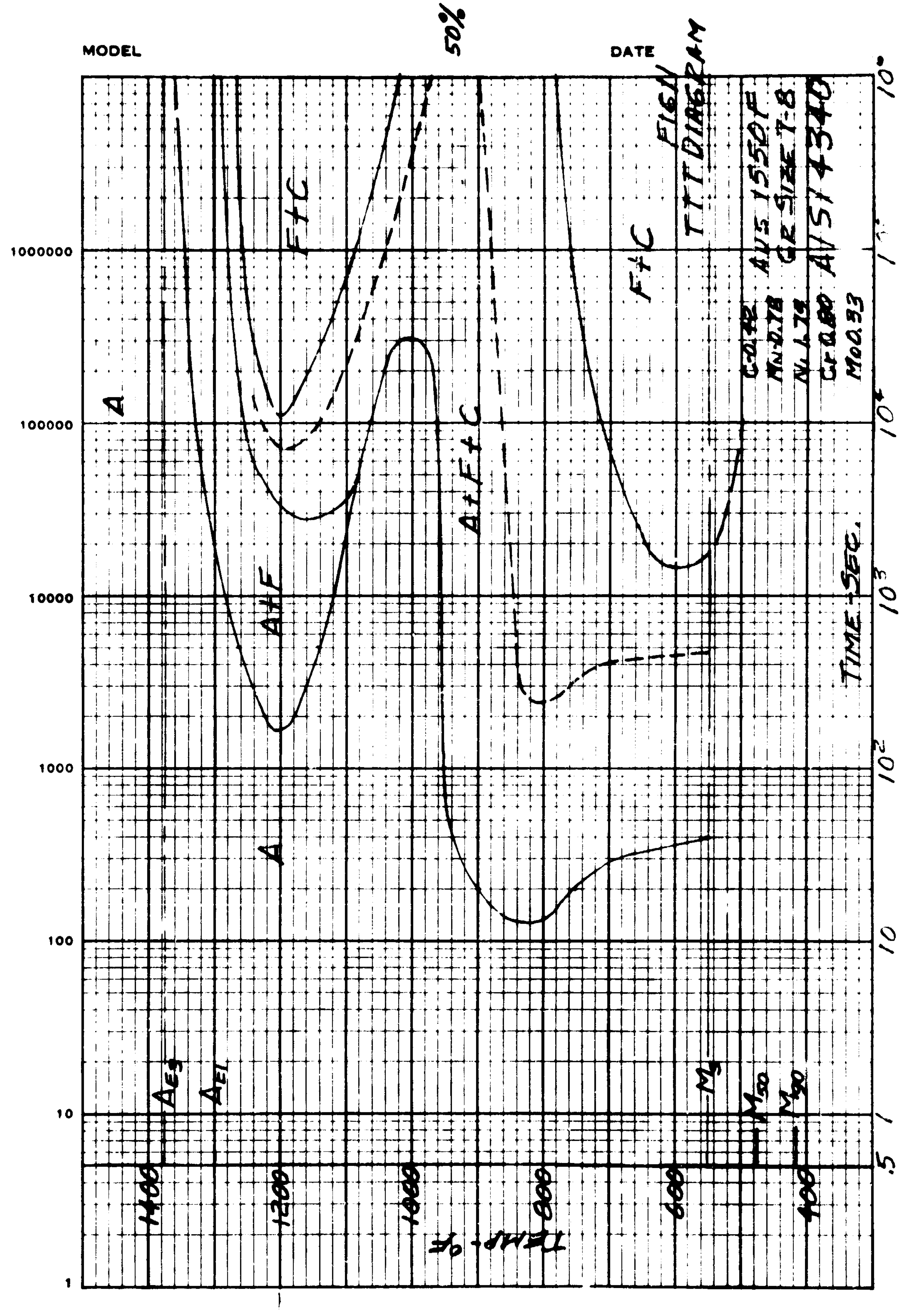

MODEL
DATE
TTT DIAGRAM
AUS 1625F
GR SIZE 7-8
AISI-1340
C-0.43
MN-.75B
A
A+F
A+F+C
F+C
30%
A_{e3}
A_{e1}
M_s^*
M_{50}
M_{90}
TEMP-°F
1400
1200
1000
800
600
400
TIME-SEC
1M
1H
1D
1W
1
5
10
10^2
10^3
10^4
10^5
10^6
1
10
100
1000
10000
100000
1000000

FIG 8


MODEL
DATE
FIG 9
TTT DIAGRAM
AUS 1450F
GR SIZE 7-8
AISI 2340
C: 0.37
Mn: 0.68
Ni: 3.41
1000000
100000
10000
1000
100
10
1
10^6
10^5
10^4
10^3
10^2
10
1
.5
TIME - SEC
A
A+F
F+C
A+F+C
90%
A_{es}
A_e
M_s
M_{50}
M_{90}
1400
1200
1000
800
600
400
TEMP - °F


MODEL
DATE
FIG 10
TTT DIAGRAM
AUS 1550F
GR SIZE 7-8
AISI 4140
C-0.37
Mn 0.77
Cr 0.98
Mo 0.21
TIME SEC
1000000
100000
10000
1000
100
10
1
10^6
10^5
10^4
10^3
10^2
10
1
.5
A
A+F
A+F+C
F+C
50%
Ae3
Ae1
Ms
M50
M90
1400
1200
1000
800
600
400
TEMP-°F


MODEL
DATE
FIG 11
TTT DIAGRAM
AUS 1550°F
GR SIZE 7-8
AISI 4340
C-0.42
Mn-0.78
Ni-1.79
Cr-0.80
Mo-0.33
50%
F+C
F+C
A+F+C
A+F
A
A
A_{es}
A_{el}
M_s
M_{50}
M_{90}
1400
1200
1000
800
600
400
TEMP-°F
TIME-SEC.
1000000
100000
10000
1000
100
10
1
10^5
10^4
10^3
10^2
10
1
.5

MODEL

DATE

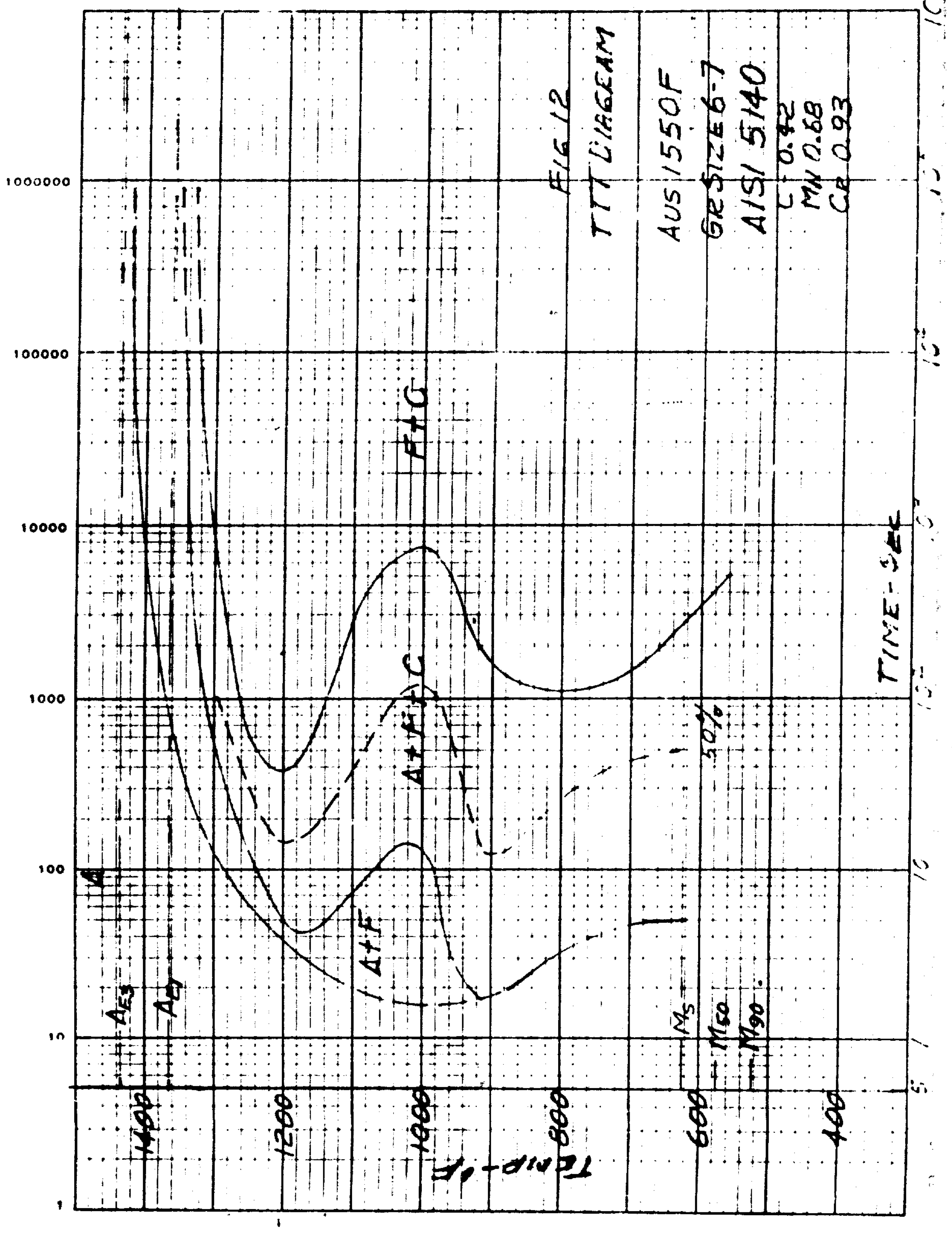


FIG 12
TTT DIAGRAM
AUS 1550F
GR SIZE 6-7
AISI 5140
C 0.42
MN 0.68
CR 0.93
F+C
A+F+C
A+F
50%
A
A_{E3}
A_{E1}
M_S
M_{50}
M_{90}
TEMP-°F
TIME-SEC
1000000
100000
10000
1000
100
10
1
1400
1200
1000
800
600
400

RATES OF DIFFUSION FOR ALLOYING ELEMENTS AT 2200 F

Element	D-Cm^2/SEC
Silicon (2370 F)	1.5×10^{-9}
Chromium	1.4×10^{-9}
Molybdenum	9.0×10^{-10}
Manganese	1.36×10^{-10}
Nickel	3.75×10^{-11}
Iron (Self Diffusion)	3.56×10^{-11}

When found, the banding in the 1340, 4140 and 5140 was more marked, while the ferrite and pearlite bands in 2340 appeared more diffuse. It must be remembered that the heat treatment used in these tests was intended to produce the maximum variation in structure. Had a simple austenitize quench and temper heat treatment been used, the banding would have appeared as a less marked difference in the microstructure of the tempered martensite.

The 4340 steel was homogenized for 5, 16, 25, 50, 75, 100 and 200 hours, then normalized to refine the grain size in the same manner as the other steels. The degree of homogenization, as evaluated by the martensite or direct quench technique and ferrite transformation technique, was judged to be complete between 5 and 16 hours at 2200 F. However, using the pearlite transformation there was some banding after 200 hours of homogenization.

Referring to Figure 11, note that transforming at 1200 F the first field encountered is A+F. In this field ferrite is being rejected from the austenite. On quenching from this region the remainder of the untransformed austenite transforms to martensite. Continuing transformation at 1200 F, the next field to be encountered is A+F+C. In this field the austenite is transformed to ferrite plus cementite (iron carbide) as platelets within the grains called pearlite.

The start of transformation and end of transformation for the five steels are shown by Figures 13 and 14. Note the effect of the alloying elements on the shape of the TTT curves of the five steels. Consider the other steels in relation to the 1340 steel. Chromium and molybdenum in 5140 and 4140 cause the curves to be shifted to the right requiring longer times to start the transformation of austenite at 1200 F. Above the nose of the curve the temperature of complete transformation from austenite is very close for 4140, 5140 and 1340, Figure 14. Nickel, as in 2340, has little effect on the time, but the temperature is lowered considerably. The combined effect of the nickel, chromium, and molybdenum is to lower temperature and increase time as shown by 4340. They produce the "bay" between 800 and 1000 F in which austenite will remain stable for 3×10^4 seconds or 8 hours.

At 1200 F an area of the microstructure, lean in any or all of the alloying elements of a 4340 steel, would start to transform faster than a richer area. In addition, the transformation would be completed sooner. The only steel for which transformation at 1200 F would not be complete in the range of the diagrams is 2340. Referring to Figure 9, this transformation remains in the A+F region for the entire time range of the diagram.

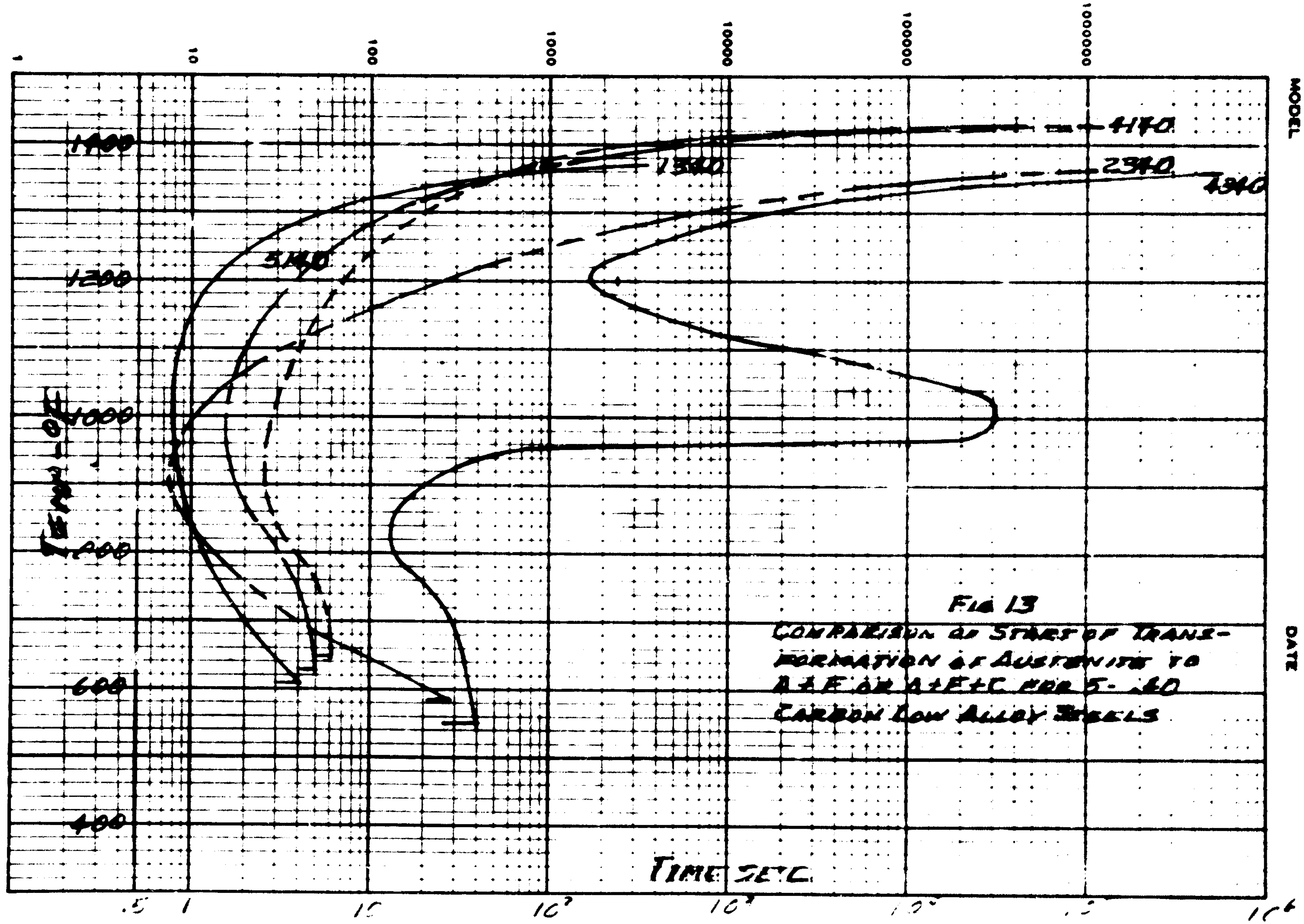
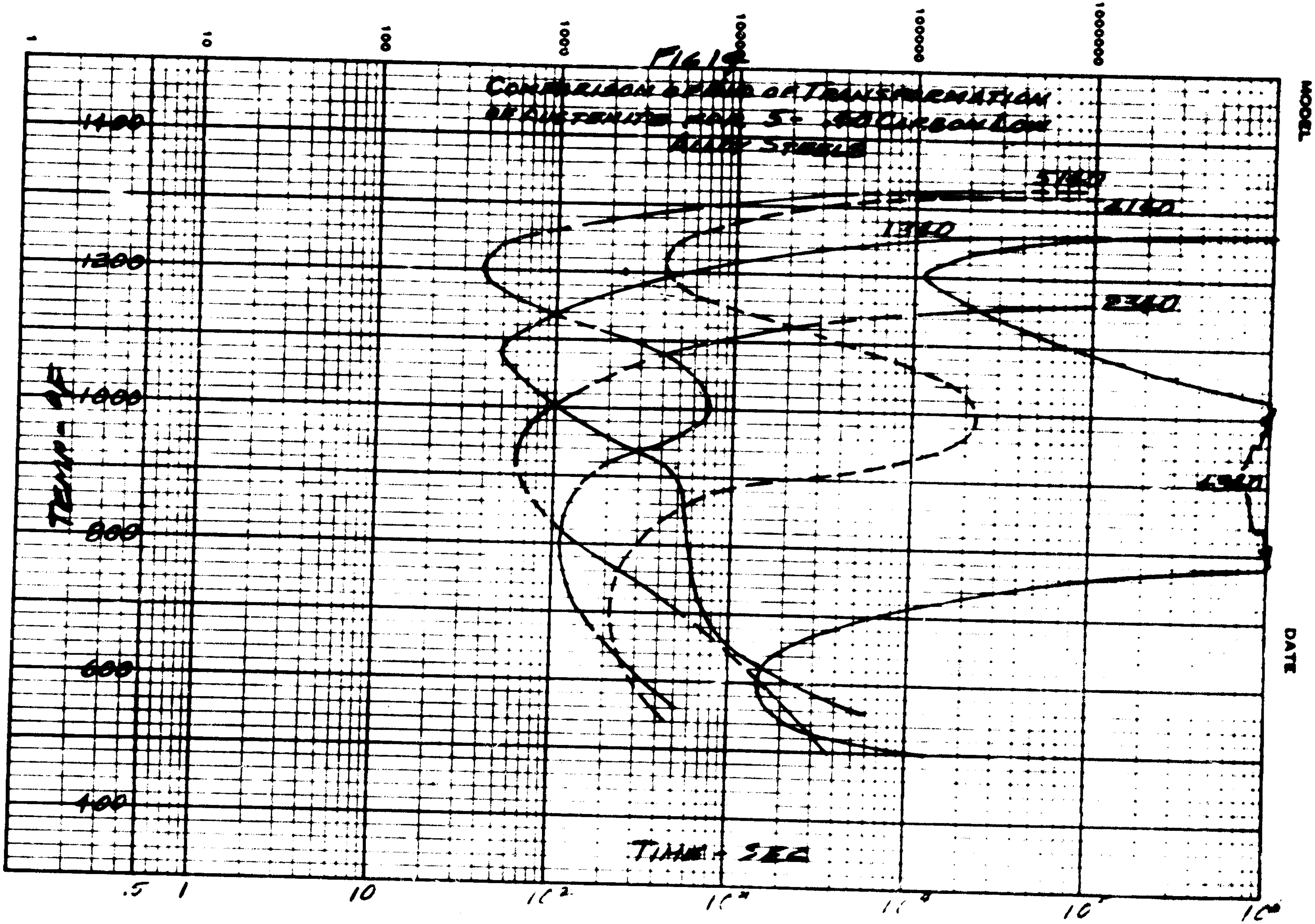


FIG 13
COMPARISON OF START OF TRANSFORMATION OF AUSTENITE TO A+F OR A+F+C FOR 5-.40 CARBON LOW ALLOY STEELS

FIG 16

COMPARISON OF SPEED OF TRANSFORMATION OF AUSTENITE FOR 5- .40 CARBON LOW ALLOY STEELS



That the transformation is never complete indicates that the transformation is very sluggish. It also indicates that nickel is the cause of the banding as revealed by the pearlite technique even though pearlite is not formed by a 2340 steel at 1200 F. While there is not complete segregation of the alloying elements, the effect of nickel enrichment can only be described as a tendency to shift the TTT diagram toward the characteristic of the steel using that element alone. This shift and the slow diffusion rate of nickel as shown by the non-homogeneity of 2340 steel, and the complete homogeneity of 1340, 4140, 4340 and 5140 steels, all point toward nickel as the cause of banding.

Tensile tests were done on "as received" and homogenized 4340 in the normalized condition. The strengths did not vary significantly, but the elongation was increased from 18.6 to 20.6% in the longitudinal direction and from 12.8 to 15.3% in the transverse direction. The charpy impact strength was increased from 82 to 86 ft. - lbs. in the longitudinal direction and from 31.4 to 39 ft. - lbs. in the transverse direction. The authors concluded that homogenization produced an improvement in transverse ductility, but that the degree of improvement was not commercially significant.

A considerable amount of work was done by Wyss(19) in Switzerland on low alloy steels. Castings were hot worked by three amounts, 15:1, 100:1 and 650:1. The samples were austenitized and isothermally transformed at 1270 F. The primary band spacing was inversely proportional to the square root of the degree of deformation. The banding was found to disappear as the band spacing approached the grain size of the steel. Further work(20) reported later indicated that for unalloyed steels phosphorous segregated areas favored ferrite formation. In alloyed steels, he reported that ferrite formed in alloy-poor dendrite areas from the original casting. He reported that fine grain size, greater amounts of working and low segregation in the original casting favored unbanded steel, while the opposite favored banding.

In work done in Germany about the same time, Plockinger(21) worked with plain carbon and Cr-Mo steels. By diffusion annealing (temperature not reported in abstract, but presumably in 2000 F range) for 30 hours the plain carbon and alloyed steel were homogenized. He reported that chromium and molybdenum had a strong influence on banding.

In the latest work found(22), three steels were cast under conditions to give fully columnar, partially columnar and equiaxed. and fully equiaxed structures. The steel used was a 1-1/2% Ni-Cr-Mo, .30-.35 carbon steel which is similar to 4330 or 4335 under AISI designation. The ingots were forged and samples were examined as cast, as forged, and as heat treated by austenitizing and isothermally transforming at 1180 F for 15 minutes, then quenching. Electron microprobe analyses were made of the structures.

It was found that nickel and chromium segregated much more in the equiaxed areas of the three castings than in the columnar areas. Banding developed more severely in equiaxed zones of the castings. Forging which reduced segregation was more effective in reducing the chromium segregation than the nickel.

CONCLUSIONS

1. Banding is caused by segregation of the alloying elements in ingots during solidification.

2. During subsequent hot working operations the segregation is aligned in the direction of working and results in a banded appearance of the microstructure.

3. The diffusion rates of the alloying elements in steel control the homogenization of the ingot, chromium and molybdenum homogenizing readily and nickel very slowly.

4. The increase in properties by homogenizing does not warrant the involved process of homogenizing, then refining the resulting large grain size.

5. The fact that design can orient the plane of maximum stress in a direction parallel to the working direction of the stock or forging, it is felt, has lead to the economic infeasibility of performing the very involved and costly task of homogenization. This behoves the designer to take as much advantage of controlling grain flow as possible. ([illegible] AD-[illegible])

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