SOME OF THE EFFECTS OF BAINITE ON THE IMPACT PROPERTIES OF STEEL

Thesis for the Degree of M. S.
MICHIGAN STATE UNIVERSITY
Herman Eugene Day, Jr.
1964

ABSTRACT

SOME OF THE EFFECTS OF BAINITE ON THE IMPACT PROPERTIES OF STEEL

by Herman Eugene Day Jr.

An investigation was conducted to verify the superiority of lower bainite over tempered martensite as well as to determine the effects of small amounts of upper bainite on the transition curve for SAE 4340 steel.

Charpy V-notch specimens were prepared from aircraft quality, cold-rolled stock. These specimens were then heat treated to develop microstructures varying from oil-quenched martensitic to fully bainitic in the lower bainite region. Transformation in the upper bainite region was conducted at 825° F and times of transformation taken every ten seconds up to a maximum of three minutes. Care was taken to temper all specimens to nearly identical strength levels.

It was found that throughout the temperature range of the test, a fully bainitic structure was superior to the quenched and tempered one. No explanation other than the one offered by E. C. Bain in 1934 is available (6). His theory involved the presence of sub-microscopic cracks in the quenched and tempered structure.

The effect of upper bainite on the transition curve was marked and involved a lowering of the impact strengths as well as a shifting of the transition temperature to higher values. Although no precise

estimate of the percentage of upper bainite was made, the time of transformation required to produce the critical amount was narrowed to a ten second interval between 60 and 70 seconds at 825° F. Examination of the untempered microstructures indicated that this time also coincided with the production of a continuous network of bainite and this fact was used to explain the deleterious effect of upper bainite on the toughness of SAE 4340.

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By

Herman Eugene Day, Jr.

A THESIS

Submitted to
Michigan State University
in partial fulfillment of the requirements
for the degree of

MASTER OF SCIENCE

Department of Metallurgy, Mechanics, and Materials Science

1964

ACKNOWLEDGEMENT

The author is deeply indebted to Dr. Howard L. Womochel for his guidance during the course of this work. His sincere interest and many thought provoking discussions were of inestimable value. The author is also grateful to Dr. A. J. Smith, Professor and Head of the Department of Metallurgy, Mechanics, and Materials Science, for his frequent counsel and advice. The Division of Engineering Research deserves much credit for the speedy and accurate preparation of the numerous specimens used in this research. Finally, a debt of gratitude is owed to the Department of the Army and the Ordnance Corps who sponsored me during my gratuate studies and research.

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INTRODUCTION

Although tempered martensite is generally considered to be the most desirable end-product in the heat treatment of steel, it is not uncommon in commercial practice to encounter slack quenched structures containing varying percentages of either upper or lower bainite or pearlite. The identity and the amount of the alien microconstituents is a function of the temperature of formation and the alloy content of the steel. The latter factor determines the characteristics of the isothermal transformation diagram.

In low alloy or plain carbon steels it is extremely difficult, if not impossible, to cool a significant section at a rate great enough to avoid an as-quenched microstructure containing high-temperature transformation products. A characteristic isothermal transformation diagram for such a steel is shown in Figure 1a. When steels are more highly alloyed, it is possible to shift the pearlite nose of the diagram to the right, thereby facilitating quenching of the steel to a fully martensitic structure. This effect is as shown in Figure 1b. The critical cooling rate, that is the cooling rate which must be achieved in order to insure full hardening of the specimen is as shown in red. Note that in Figure 1a, this rate is severe whereas in Figures 1b and 1c, it is much less severe and therefore, more easily attainable in significant sections using standard quenching techniques.

The quenched structure, when suitably tempered, will normally give the best combination of strength, ductility, and toughness. Further alloy additions may generate a bainitic shelf as shown in Figure 1c. This type of diagram characterizes SAE 4340 and other triple alloy steels containing Cr, Ni, and Mo.

In a steel which has an isothermal transformation diagram characterized by a "bainite knee," the pearlite nose is pushed quite far to the right. In this case it is the lower or "bainite knee" which must be avoided if the specimen is to be fully hardened. It can further be seen that transformation in the region of the knee as well as just above it, is very slow and difficult to bring to any degree of completion. Only in the range of temperatures just above M_s is it practical to achieve 100% bainite in such steels. The data below which are typical for SAE 4340 steel are given to demonstrate this point. (24)

Pearlite nose	1175° F	
M_s	570° F	
1st bainite	1020° F	
Highest all bainite	610° F	
Time for bainite completion at	M_s	2.7×10^3 sec
	$M_s + 20$	4.7×10^3 sec
	$M_s + 50$	over 10^5 sec

The difficulty inherent in producing 100% bainitic structures throughout the temperature range of bainite formation accounts, in large part, for much of the confusion and lack of data concerning this structure so commonly found in steels. The relative ease of avoiding the pearlite nose in steels of this type, however, has made SAE 4340 a standard for experimental work on bainite. This, coupled with the fact that SAE 4340 provides the base composition for many of the ultra-high strength steels, justifies its use in this investigation.

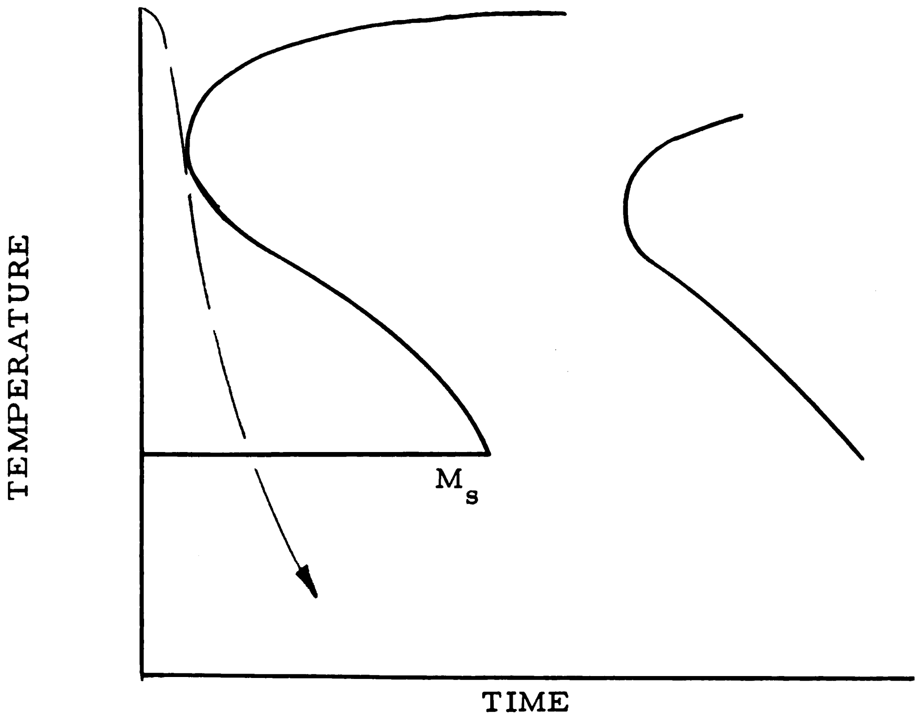

Figure 1a - Characteristic T - T - T curve for a plain carbon or low alloy steel.

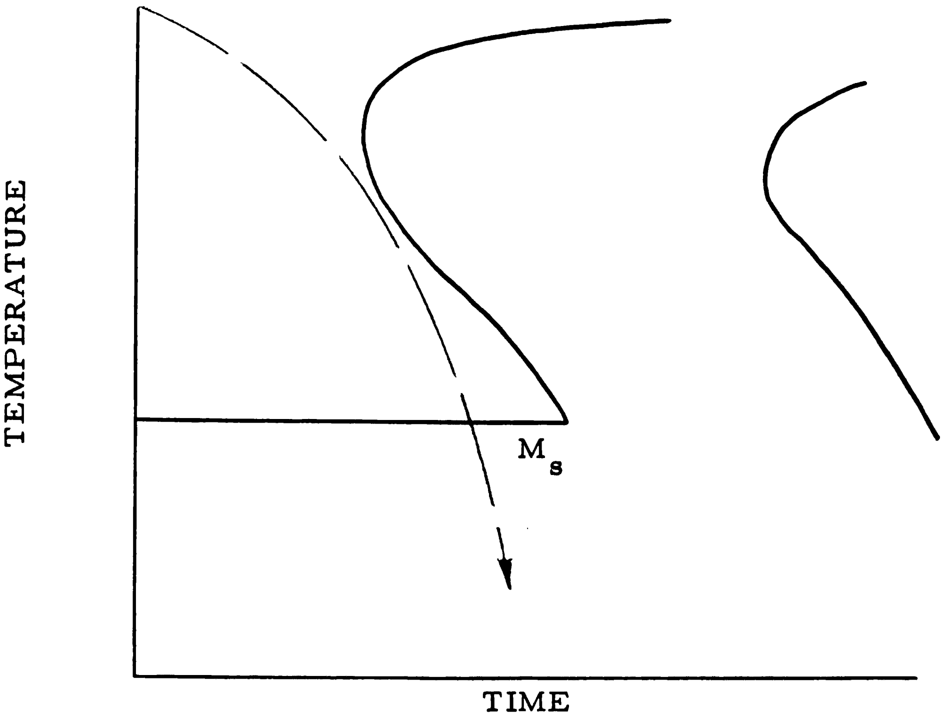

Figure 1b - Effect of alloying elements in shifting pearlite nose to longer transformation times.

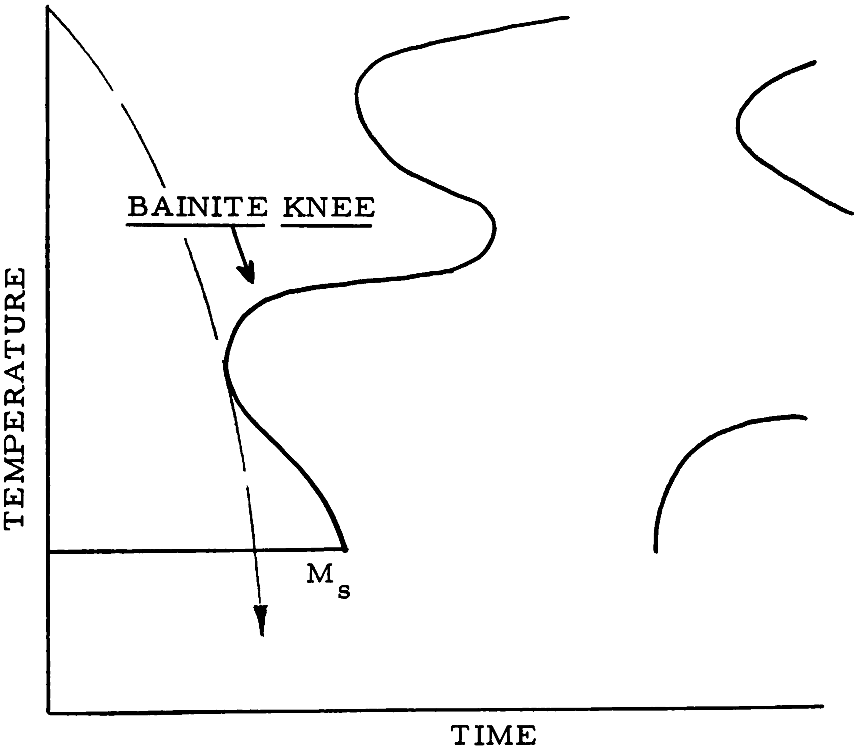

Figure 1c - Effect of alloying elements in producing a bainite "knee".

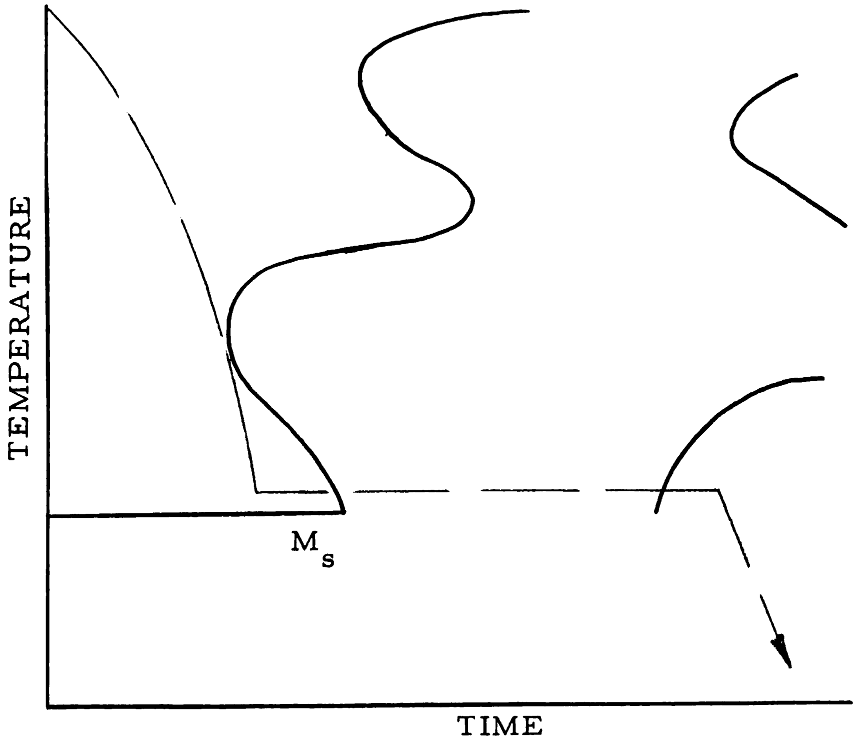

Figure 2 - Typical cooling cycle for austempering steel.

In several investigations (4-8) it has been demonstrated that isothermal transformation at low temperatures slightly above M_s has resulted in structures having high hardness with ductility and toughness properties superior to those of tempered martensite at the same strength level. This heat treating procedure is commonly referred to as austempering. A typical cooling cycle superimposed upon an isothermal transformation diagram is as shown in Figure 2. Other investigations have shown a deterioration in these properties after decomposition in the bainite range. (9-15) In the cases where good properties have been obtained, it is probable that the experimental heat treatment resulted in homogeneous microstructures of 100% bainite. The cases where poor properties resulted certainly involved heterogeneous microstructures containing bainite, martensite, and possibly other transformation products. A standard quench and temper cycle is as shown in Figure 3.

There is confusion associated with the results of austempering at high temperatures in the bainite range resulting in partial transformation to bainite followed by quenching and tempering to intermediate hardness levels. Lack of accurate metallographic data has made it impossible to isolate the influence of

1) Percent of bainite
2) Temperature of transformation
3) Subsequent tempering treatment

on the mechanical properties of steel. (3)

Particular interest has centered around the influence of admixtures of bainite in tempered martensite on the transition temperatures of various steels. One of the peculiar characteristics of steel is the tendency for the material to become brittle at low temperatures. Initial interest in this

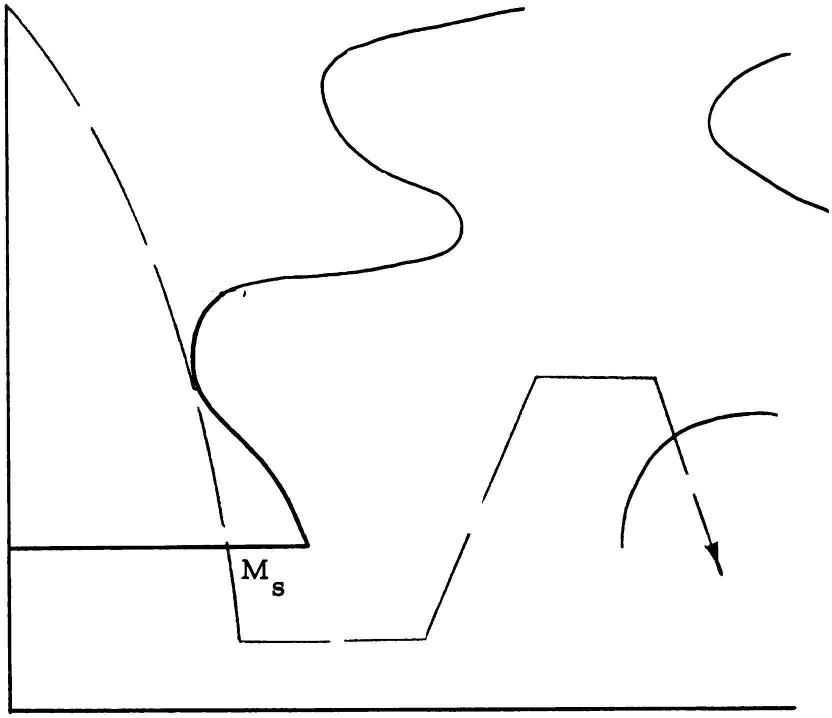

Figure 3 - Standard quench and temper cycle.

area developed when a series of catastrophic failures of ships, bridges, and pressure vessels were noted and no explanation was readily available for their occurrence. At least one factor was common to all of them and that was their occurrence at low temperatures. M. E. Shank (16) and K. K. Cowart (21), in a series of investigations, attempted to correlate the factors which were similar in each instance. Considerable activity in this field resulted in findings as to the effects of various metallurgical and mechanical factors on the susceptibility of steel structures to low temperature embrittlement. An excellent summary of these findings is presented in a symposium prepared by the American Society for Testing and Materials (17) and will not be elaborated upon here.

One of the principle factors in determining whether or not a steel will retain its toughness at low temperatures is its final crystalline structure. As reported by Wood et al (2) and Shank (19), face centered cubic metals do not undergo a ductile to brittle transition at low temperatures whereas all body centered cubic structures and some hexagonal close-packed structures are subject to this embrittlement. Austenitic steels, for example, retain their toughness even at very low temperatures.

Another factor appears to be the homogeneity of the final microstructure. Foley (1) and A. Hurlich (18) indicate that it is possible for steel to retain its toughness at very low temperatures if the final microstructure consists of 100% tempered martensite. The occurrence of brittle fracture and corresponding low impact energy values appears to coincide, in low and medium alloy steels, with the presence of appreciable quantities of high temperature transformation products such as bainite, pearlite, and ferrite as admixtures with tempered martensite. As previously noted, however, a 100% bainitic structure has often proved to be superior to tempered martensite throughout the temperature range for a given strength level. This fact will be confirmed in this paper. The inherent difficulty of producing fully bainitic structures in the majority of steels at any given temperature level has prevented austempering from being a universally applied technique.

In work by Hehemann et al (3), the influence of bainite on the properties of SAE 4340 at high strength levels has been investigated. In the investigation, an attempt was made to relate the effect of percent bainite present in the final microstructure, the temperature of formation of the bainite, and the transition temperatures for this steel. It appears that too many variables were involved to allow a clear separation of their effects. In the above mentioned work it is implied that the form of transition curve is controlled by the characteristics of the dominant microconstituent. In other words, it obeys the laws of mixtures. Thus, a steel composed primarily of tempered martensite should have a low transition temperature while one having primarily a bainitic structure should be characterized by a high transition temperature. It is further implied that the transition from one characteristic to the other occurs over a fairly narrow range of decomposition product.

If this is so, then it would be of interest to know the limits of this range and thereby determine an upper tolerance level for heterogeneity in a steel when the microconstituents are tempered martensite and upper bainite.

The purpose of this paper is twofold. First, an investigation was made to verify the often stated but seldom quantitatively justified conclusion that a 100% lower bainite is superior to 100% tempered martensite at the same strength level. Secondly, an attempt was made to define the critical bainite content for SAE 4340 which, if exceeded, results in serious deterioration in toughness as indicated by the transition curve. There are two purposes which may be served by the latter determination. First, we may determine the percentage of upper bainite which can be tolerated in a slack-quenched structure without seriously impairing toughness. Secondly, we may determine the amount of upper bainite which may intentionally be formed as in the bainitic hardening of steel. In work by Baer and Payson (7) it has been shown that partial transformation of a high speed steel at high temperatures in the bainite range results in retardation of the transformation at lower temperatures and therefore provides a significant improvement in hardenability with only a slight loss in overall hardness. This stabilization of the bainite reaction is further demonstrated in work by Hehemann and Troiano (20). In the latter investigation, high temperature transformation was conducted near B_s in SAE 4340 steel with the allowed time of reaction being sufficient to cause the transformation to go to completion for all practical purposes. It was shown that this treatment resulted in a new induction period before the bainite reaction recommenced at lower temperatures. This retardation of the bainite reaction at lower temperatures accounted for an appreciable increase in the hardenability due to the stabilization of the austenite which was subsequently transformed to secondary bainite during tempering. Again there was only a nominal loss in overall hardness. No attempt was made in this article, however, to evaluate the properties of the steel so heat treated.

In subsequent work, upon which this paper is based, Hehemann has shown that there is a serious diminution of mechanical properties if too much high temperature transformation product is present in the microstructure. Thus, it is hoped that this investigation will shed some light on the effectiveness of bainitic hardening by establishing the maximum allowable high temperature transformation which may occur without impairing the toughness of the final structure.

Perhaps it would be helpful to discuss briefly the nature of this transformation product called bainite which has been the subject of so much controversy. Basically, bainite is a dispersion of iron carbide particles in alpha ferrite. In SAE 4340 it may occur after prolonged holding in the range of temperatures from 1020° F to below M_s. If SAE 4340 is quenched to a temperature below 1020°F but above M_s and then held for a period of time, bainite will begin to form by a process of diffusion of carbides. In the upper portion of the bainite range it is virtually impossible to transform the steel to 100% bainite due to the self stopping nature of the reaction. It has been reported that in this steel, 100% bainite cannot be obtained at temperatures above 610°F. (24) This is doubtful although at higher temperatures, the time for transformation would be excessive. If a specimen is quenched to some temperature below M_s but not low enough to achieve a 100% martensite transformation and then held at temperature for a sufficiently long time, the resulting structure will be tempered martensite and lower bainite.

There is considerable evidence to support the contention that there are two distinct forms of bainite referred to as upper and lower bainite depending upon the temperature of formation. Work by Hehemann and Matas (25) has indicated that the transition between the two forms occurs in a narrow range of temperature in the vicinity of 650° F. No explanation has been offered for this transition although there are basic morphological differences between the two forms. Under the optical microscope, upper bainite may appear feathery whereas lower bainite is more acicular in appearance.

Recent work by several investigators has shown that upper bainite is generally in the form of rods of approximately rectangular cross-section. The carbides in this form are oriented parallel to the growth direction of the bainite rods. In lower bainite, the structure is truly plate-like and the carbides occur as cross-striations oriented approximately 55-60 degrees from the direction of growth.

It is interesting to note the similarities between low and high carbon martensites and upper and lower bainites. Both low carbon martensite and upper bainite form as sheets of closely spaced needles on $\{111\}_\gamma$ planes while higher carbon martensites and lower bainite form as plates with $\{225\}_\gamma$ or other orientation habits. This analogy between lower bainite and high carbon martensite was made early by T. Ko and

S.A. Cottrell. (27) In their work, they explained that the bainite nucleus will grow if the thermodynamic requirements are fulfilled by a continuous removal of carbon from the bainite structure and by the condition that the coherency of the bainite and austenite lattices is maintained. It was proposed that the removal of carbon could be accomplished either by partitioning to the austenite or by precipitation as a carbide. At higher temperatures, where the diffusion rate is great, it is reasonable to expect the depletion of the carbon in the bainite to occur by its partitioning to the surrounding austenite. At the temperatures of lower bainite formation where diffusion is relatively slow, the logical mode of rejection would be in the form of a carbide precipitate. Thus, lower bainite would remain relatively rich in carbon.

The exact nature of the carbide in bainite depends upon the temperature of formation. In upper bainite there is no question that the carbide phase is cementite. There is less certainty as to the nature of the carbide in lower bainite. Magnetic measurements reveal no cementite Curie point and dilatometric studies produce total expansions which are larger than those predicted on the basis of a ferrite plus cementite structure. (25) These effects have been attributed to the recently identified epsilon carbide which is found in bainite formed below 400°F.

The first carbide to form in lower bainite is obviously epsilon, however, this is replaced by cementite after prolonged holding at temperature. At high transformation temperatures in the lower bainite region, the transformation to cementite takes place in the time required for the growth of the individual bainite plates so epsilon carbide is not normally present in the final microstructure. The precipitation of epsilon carbide and its subsequent transformation to cementite is analogous to the first and third stages of tempering.

In upper bainite, cementite precipitates directly and only in silicon treated steels can epsilon carbide be retained at temperatures as high as 600°F.

The use of the impact test in the testing of materials is of special significance. It has been shown that although steels may demonstrate excellent ductility and high strength during room temperature tensile testing, they may become very brittle when subjected to either impact loading or low temperatures. (22) It has further been stated that very high rates of loading, such as when a structure is struck by a projectile, are equivalent to very low temperatures where embrittlement is concerned.

The transition from ductile to brittle fracture may be evidenced by the relative energies absorbed during the breaking of the specimen as well as by the characteristics of the fracture itself. For military and industrial purposes, the transition from ductile to brittle fracture is often taken as occurring at the point where impact energy falls below 15-20 ft. lbs. Specimens having impact strengths greater than this figure are considered to be ductile and those having less are brittle. If temperature is the variable during the test, then the temperature at which this impact energy level occurs is called the transition temperature. (23) Some investigators have chosen to use the temperature at which the curve of impact values first deviates from the energy level characteristic of the super-transition plateau as the transition temperature. In this investigation, the latter definition will be used although the other criterion will not be ignored since, for many purposes, the low temperature tail of the curve will have great significance. Such purposes might include establishing minimum acceptance levels for items of equipment which may be put into service under conditions which would subject them to wide variations in temperature. Construction equipment, ordnance items, and aircraft parts are but a few categories of such material.

The appearance of the fracture surface itself has also been used to define the transition from ductile to brittle fracture by establishing the ratio of crystalline to fibrous fracture areas. This technique was not used herein although fracture appearance was noted and remarks will be made if appropriate.

MATERIALS AND PROCEDURE

The material used was SAE 4340 cold rolled aircraft quality steel obtained in three-quarter inch round bars. All surfaces involved in the fracture were left .015'' oversize to allow the removal of any decarburized surface during final finishing.

Chemical analysis of each of the bars was as follows:

Bar	C	Cr	Ni	Mn	Mo	Si
D 11	.40	.77	1.75	N/A	.23	.013
D 12	.38	.80	1.83	.82	.21	.014
D 13	.42	.78	1.81	N/A	.23	0.13

SERIES I

Previous investigators had resorted to a one hour normalizing treatment of all stock at 1650° F followed by a four hour anneal at 1200° F. This technique was employed on a representative sample and the resulting microstructure compared to that of the stock in the as received condition. No significant improvement was noted so this procedure was not employed. Microscopic examination of the stock yielded an apparent McQuaid-Elm grain size of 7-8 which agreed with the data provided by the supplier.

The M_s for this steel was estimated based upon the lowest carbon analysis obtained and using the equation below which was taken from the Metals Handbook, 1948 edition.

$$M_s(F) = 930-540(\%C)-60(\%Mn)-40(\%Cr)-30(\%Ni)-20(\%Si)-20(\%Mn)$$

From this equation the Ms was estimated to be 579° F.

All specimens were austenitized at 1550° F for thirty minutes in neutral salt prior to quenching or isothermal transformation. Isothermal transformation was performed in agitated salt at temperatures between B_s and B_f depending upon the desired final microstructure.

Temperatures for impact testing above room temperature were achieved in an oven in which specimens were held for thirty minutes prior to breaking. Temperatures below room temperature were achieved using crushed ice and water for 32° F and varying amounts of solid carbon dioxide in alcohol for very low temperatures. Since the sublimation point of solid CO_2 is -108.5° F, this was taken as the lowest temperature of the test. Specimens were immersed in the low temperature bath for a period of one hour prior to breaking. Tongs used for transferring the specimens from the bath to the impact machine were of special design both to facilitate rapid placement of the specimen in the machine and to minimize the heat transfer from tongs to specimen. A further precaution was taken in that the tongs were also held in the bath. The time interval between removing the specimen from the bath and breaking in the machine did not exceed five seconds and generally was less.

RESULTS AND DISCUSSION

In the first phase of the experimentation, twenty-eight specimens were austenitized as described above. Fourteen were then isothermally transformed for ninety minutes at 590° F in agitated salt and finally quenched in agitated oil. These specimens consisted of 100% lower bainite. The remaining specimens were quenched directly into oil and were thereby transformed to a fully hardened structure of martensite and a small percentage of retained austenite which is inherent in this steel. After final machining, the bainitic specimens were found to have an average hardness of Rc 49. A series of tempering experiments on the martensitic specimens indicated that tempering for one hour at 700° F should result in the same hardness. In fact, this tempering treatment resulted in a final hardness of only Rc 48.5. As will be seen, this hardness differential tends to support rather than diminish the validity of the conclusions based upon the test data. The two sets of specimens were then broken in duplicate at each of six different test temperatures ranging from -108° F to 240° F. It can be assumed that this temperature range encompasses the limit of operational temperatures to which this material would be exposed in service.

Specimens were broken in a Tinius-Olsen impact machine with the following characteristics:

Capacity	264 ft. lbs.
Striking velocity	16.5 fps
Hammer wt at striking edge	60 lbs.
Length of hammer (center of suspension to center of percussion)	900 mm

The results of this testing are shown in Figure 4.

In the second phase of the research it was necessary to determine the critical amount of upper bainite which, when present in the matrix of tempered martensite, resulted in a serious deterioration of impact properties. It was assumed that commercial interest would center around the occurrence of upper rather than lower bainite since this is the most

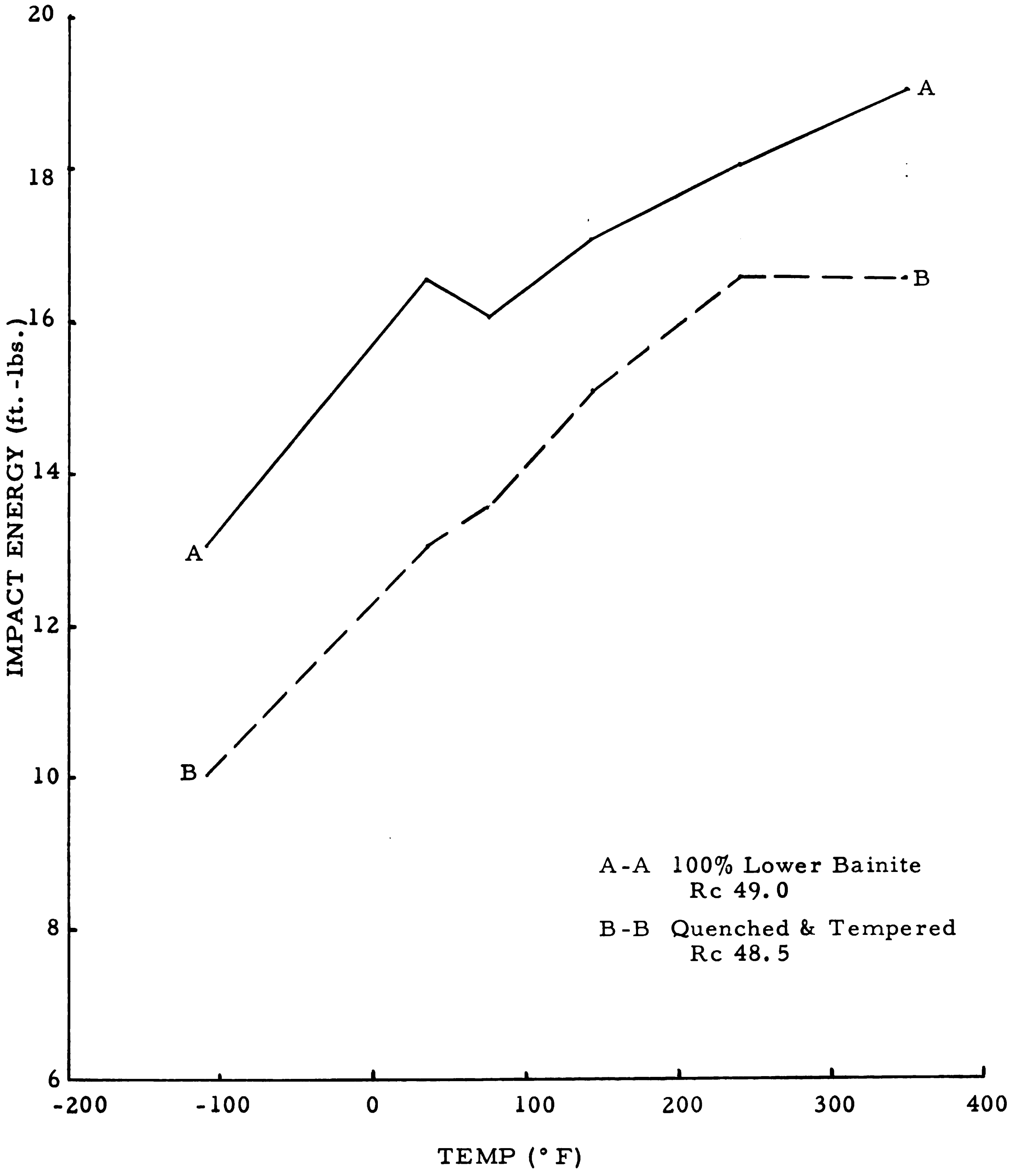

Figure 4 - Impact strengths of bainite vs. tempered martensite.

probable microconstituent to occur in slack quenched sections. It is also the partial transformation to upper bainite which affords an increase in hardenability. (7)

To determine the critical amount of bainite which can be tolerated, it is necessary to eliminate as many variables as possible. Thus, the temperature of transformation was held constant, all specimens were tempered to a uniform strength level, and only the time of transformation, which determined the amount of bainite present in the microstructure, was allowed to vary. At least one investigator (3) indicated that three minutes at 850° F resulted in 60% transformation to upper bainite. He also indicated that 15% or less upper bainite was not detrimental to the strength properties of the steel. It is significant that this amount also affords significant hardenability improvement. (20)

Thus, it was decided to limit the initial investigation to a series of specimens representing a cross-section of the isothermal transformation diagram at 825° F. Times of transformation were chosen as 15 sec., 30 sec., 1, $1\frac{1}{2}$, 2, $2\frac{1}{2}$, and 3 minutes. All of the specimens were isothermally transformed in agitated salt and quenched in agitated oil.

Four specimens were prepared at each of the reaction times concerned. Three specimens from each of the seven groups were then tempered at 800, 850, and 900° F respectively. It was hoped to achieve a complete series of specimens representing each isothermal transformation time with nearly identical hardnesses of Rc 42-43 representing a strength level of approximately 200,000 psi. The resulting tempering curves are as shown in Figure 5. Hardnesses of the finished specimens were taken as the averages of five or more readings per specimen.

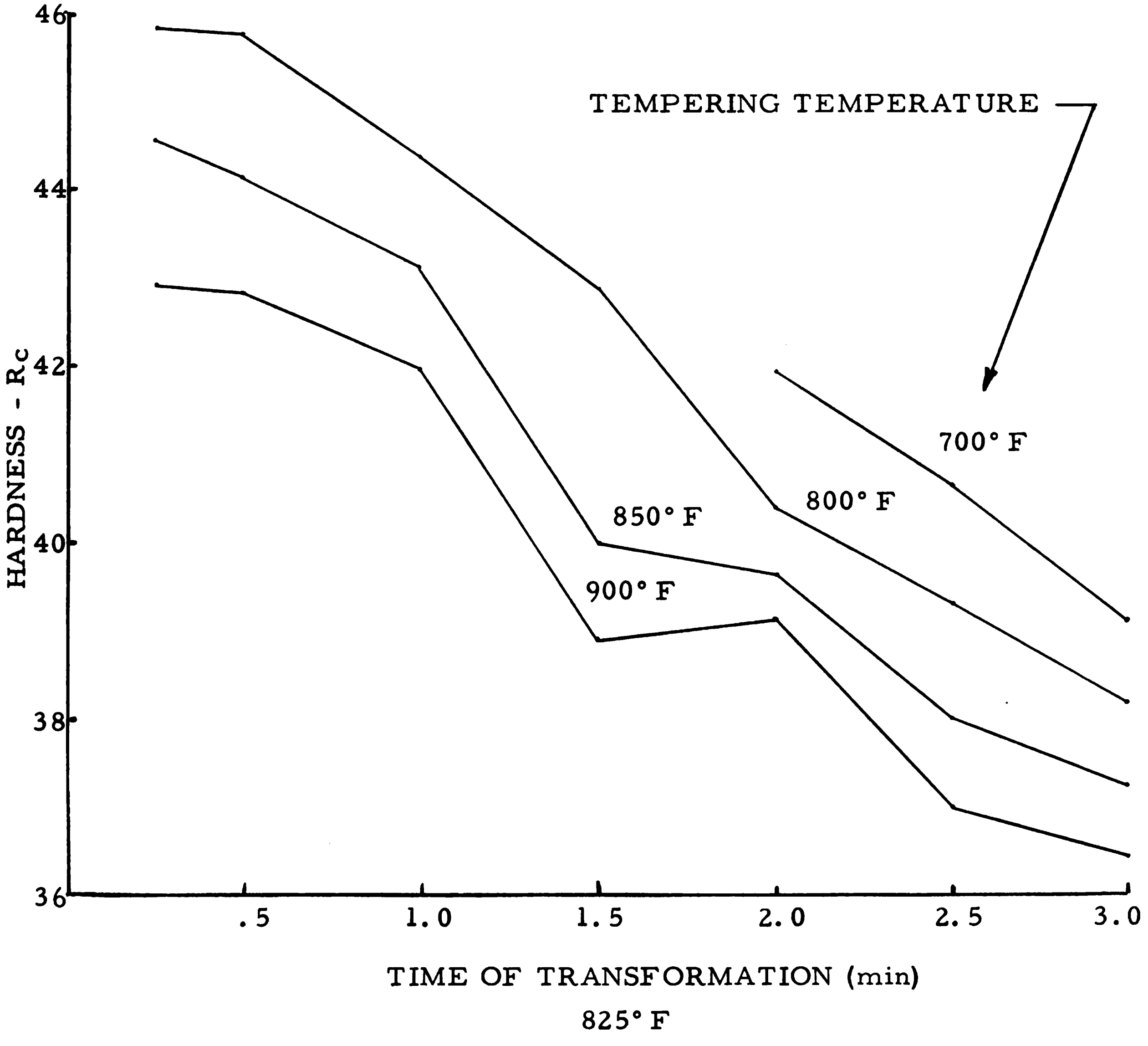

Figure 5 - Effect of tempering temperature on hardness versus time of isothermal transformation.

Four specimens were found to be of nearly identical hardness, ie. Rc 43, and were tested in the impact machine. The results are as shown in Figure 6. There is an evident decrease in toughness in the interval between one minute and one and one-half minutes. Another group of specimens with slightly higher hardness were also broken and their impact values are shown also in Figure 6. Note the similarity between the curves. Untempered samples representative of the first four time intervals were polished and etched. The microstructures appeared as shown in Figures 7, 8, 9, and 10. Note the appearance of the dark etching upper bainite in the first two photomicrographs in finely dispersed form. In Figure 9, the bainite begins to take on a more massive form and in Figure 10, the areas of bainite are quite appreciable. It is interesting to note that in Figure 8, there is an apparent continuous network of bainite beginning to form as evidenced by the region of dark-etching constituent in the center of the photograph. In Figure 9, the bainite is taking on a more massive form and is less finely dispersed and finally, in Figure 10, there is very distinct banding or continuity of the bainite structure. It is important to note that this continuous network or banding of the bainite was a characteristic of those specimens which demonstrated a high transition temperature or low impact values at room temperature.

The problem of banding in steel is a common one. It occurs principally in the direction of rolling. The photomicrographs in this paper are transverse sections of Charpy specimens, taken in the vicinity of the fracture surface. The scope of this investigation does not allow an intensive discussion of the banding phenomenon but perhaps a few comments are appropriate.

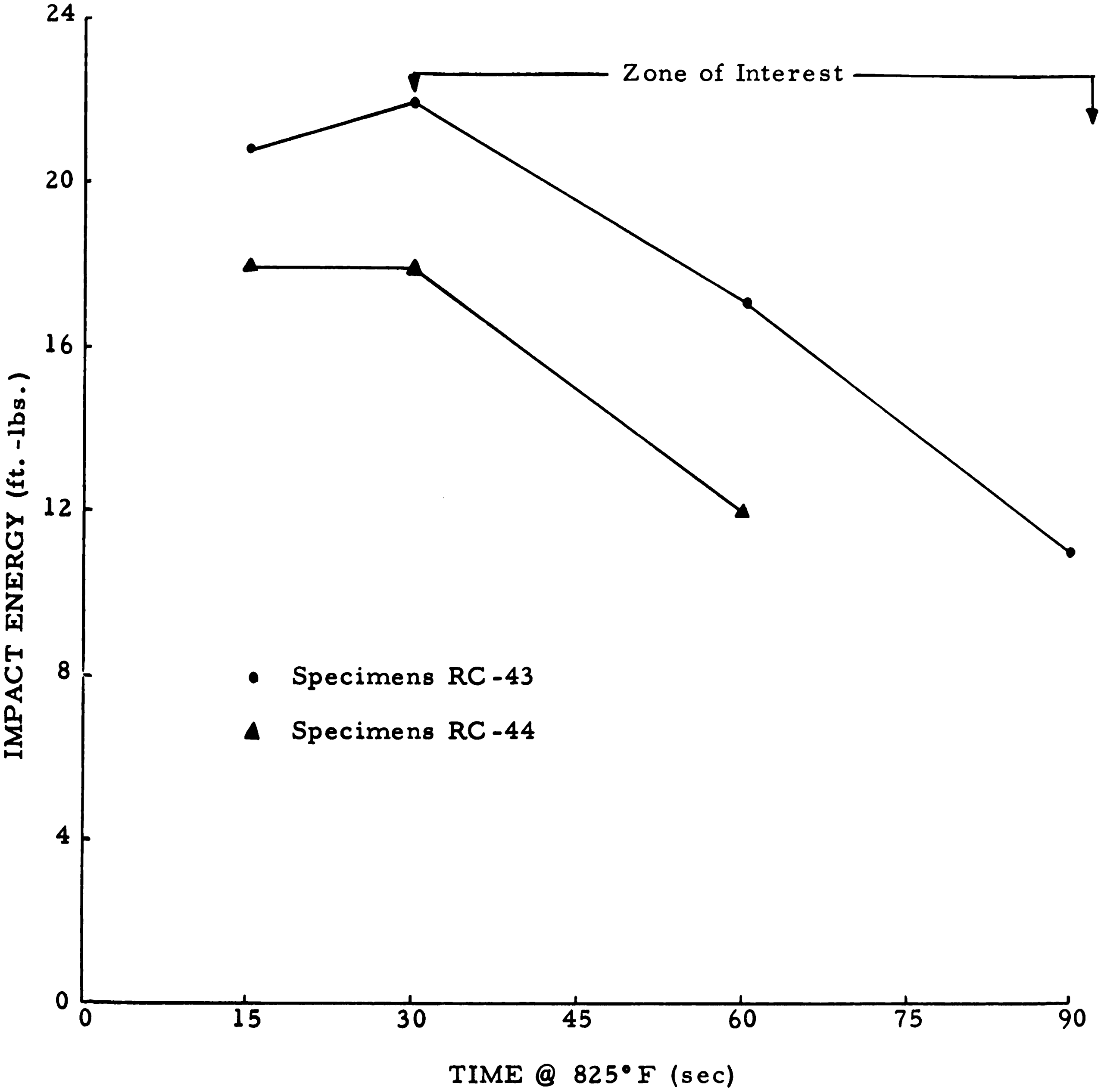

Figure 6 - Impact strength versus time of isothermal transformation. (Room temperature)

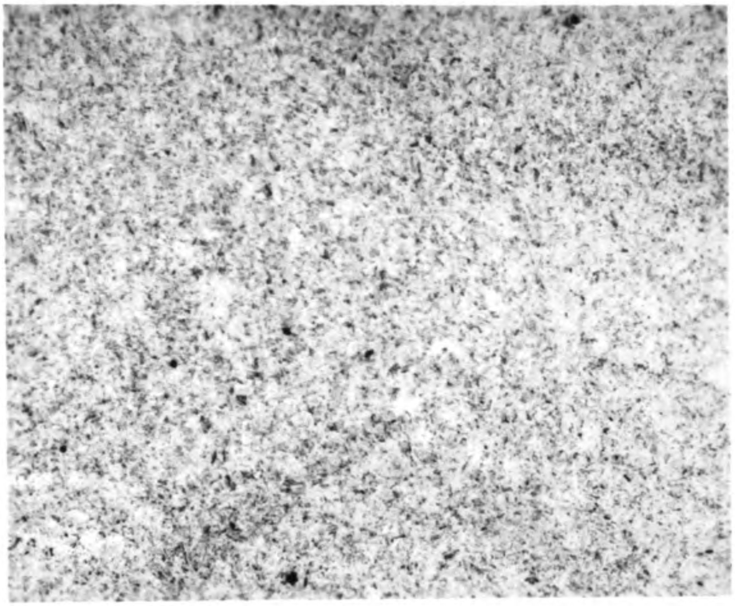

Figure 7 - Microstructure after isothermal transformation for 15 sec. at 825° F. 250x. Nital etch.

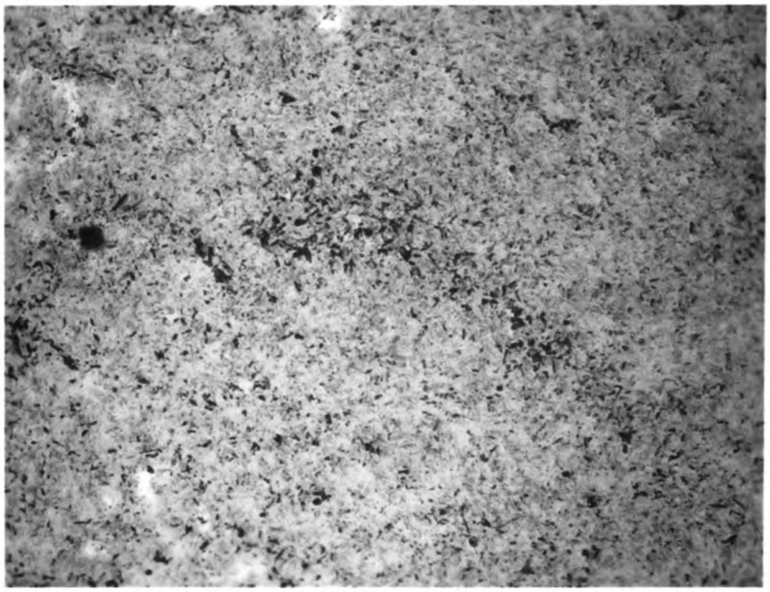

Figure 8 - Microstructure after isothermal transformation for 30 sec. at 825° F. 250x. Nital etch.

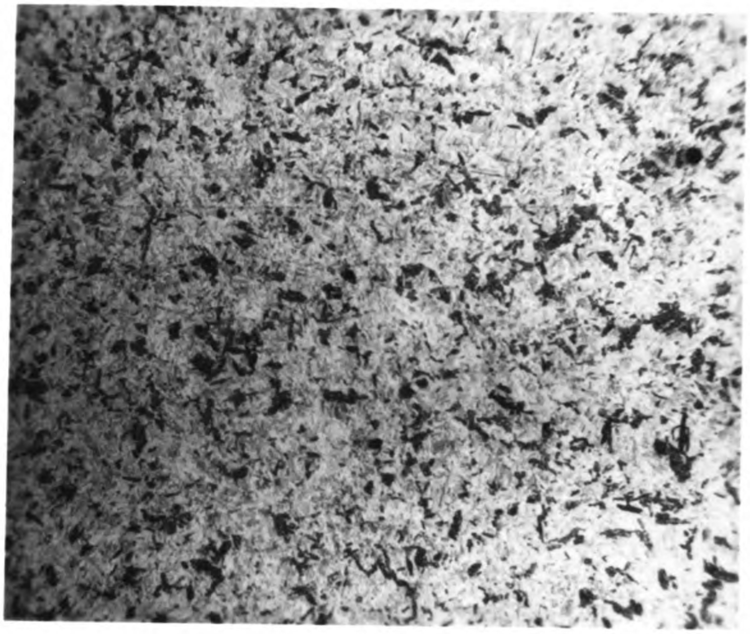

Figure 9 - Microstructure after isothermal transformation for 60 sec. at 825° F. 250x. Nital etch.

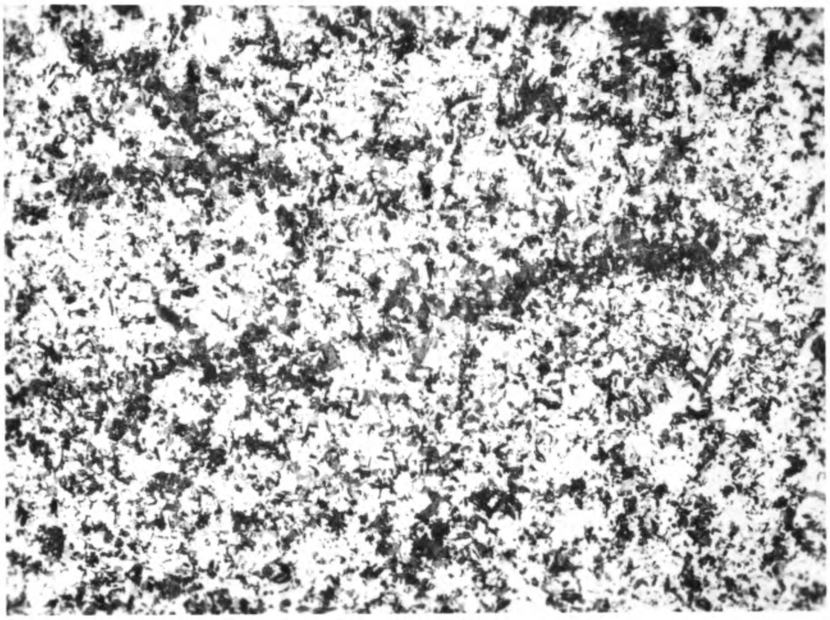

Figure 10 - Microstructure after isothermal transformation 90 sec. at 825° F. 250x. Nital etch.

As already stated, banding is frequently observed in rolled stock. If a longitudinal section is examined, layers of dark and light etching constituents will be observed. Various explanations have been offered for the occurrence of these bands and perhaps the most acceptable one is as follows. In an ingot there may be macrosegregation of both alloying elements and impurities. When the ingot is rolled, these segrated elements or phases will be extended in the direction of rolling for long distances. It has been proposed that these impurities or regions of segregation may serve as nucleation sites for ferrite. The principle effect of these bands is to impart directional properties to the steel. Impact strengths of steels have been known to vary by 50% dependent upon whether a transverse or longitudinal specimen is tested. Much work remains to be done before a firm theory of banding will be accepted. If banding is a result of nucleation of ferrite by elongated impurities or non-metallic inclusions, perhaps a vacuum melted, high purity steel would be relatively free from the effects of this phenomenon. This remains to be seen and should be investigated.

It is probable that in this investigation, the continuous network of bainite is not the product of banding but is rather a case of the bainite forming at the prior austenite grain boundaries and embrittlement occurring when the network becomes continuous. This possibility will be discussed in the analysis of results.

A plot of impact energy versus hardness is shown in Figure 11 and graphically demonstrates the importance of careful hardness control if impact energies are to be compared independent of strength level.

To confirm the initial results obtained with specimens 12-40, another series of specimens, 41-56, was selected for treatment. In

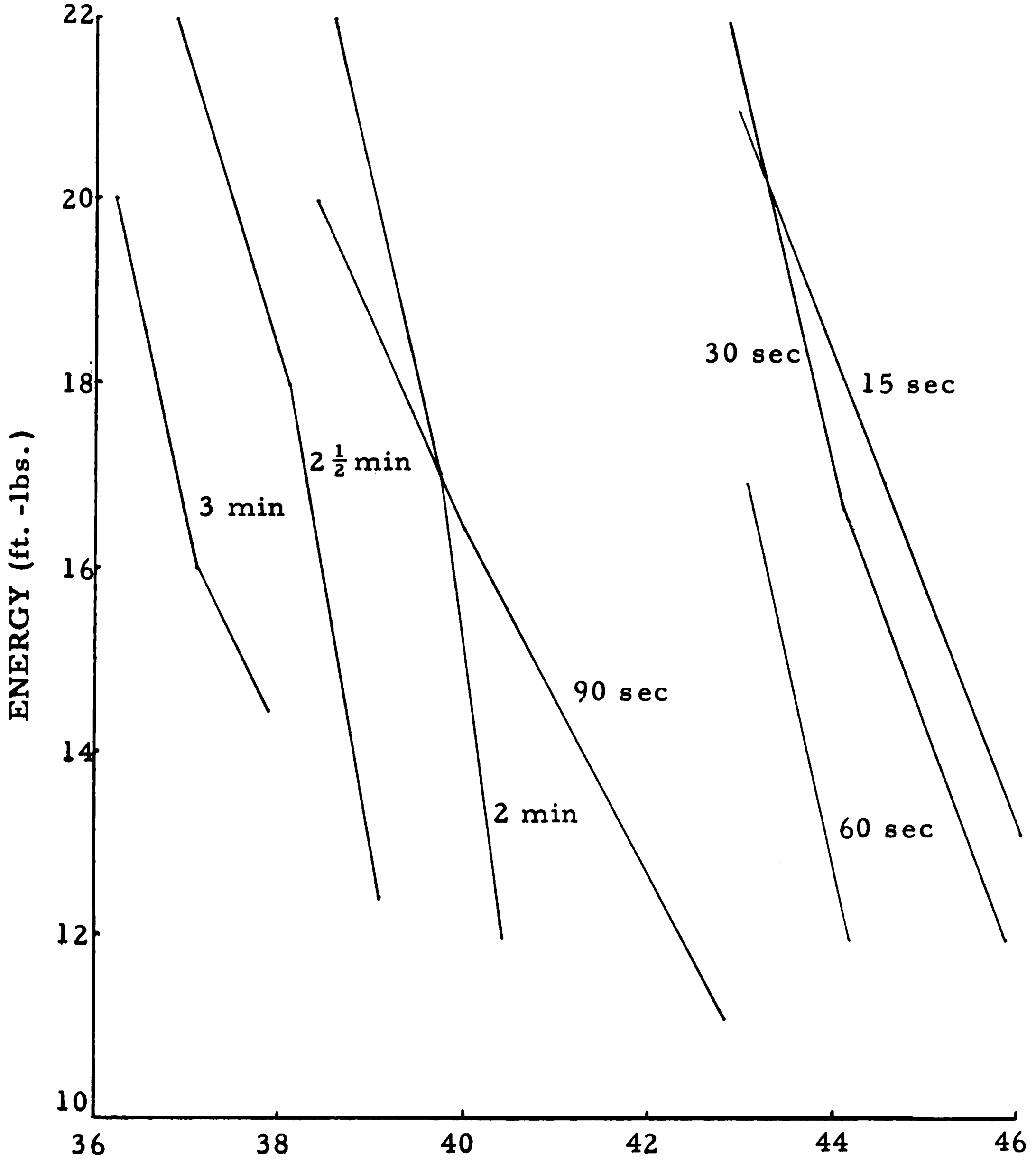

Figure 11 - Effect of hardness on impact strength.

this case, however, it was hoped to narrow down even further the range containing the critical amount of transformation product. The isothermal transformation was accomplished at 825°F for 15 sec., 30 sec., and every ten seconds thereafter up to ninety seconds.

Based upon previous tempering treatments, the following tempering cycles were chosen for each of the representative microstructures. Again all specimens were prepared in duplicate with only one set being tempered. A plot of hardness after tempering versus time of transformation is shown in Figure 12.

Specimen	Trans Time	Temper (1 hr.)	Hardness
43-44	15 sec.	900°F	41.8 Rc
41-42	30 sec.	900°F	41.5
45-46	40 sec.	875°F	41.8
47-48	50 sec.	875°F	41.7
49-50	60 sec.	850°F	41.8
51-52	70 sec.	825°F	41.1
53-54	80 sec.	800°F	41.5
55-56	90 sec.	800°F	41.2

Although somewhat lower in hardness than the first group of specimens, they were considered to be of nearly identical strength for all practical purposes. Again these specimens were broken and their impact values recorded as shown in Figure 13.

As in the previous results, there seems to be a marked deterioration in impact properties in the vicinity of one minute transformation time. To obtain an estimate of the amount of primary bainite present in the microstructure, the untempered specimens were broken and the fracture surfaces polished and etched. To obtain a correlative estimate

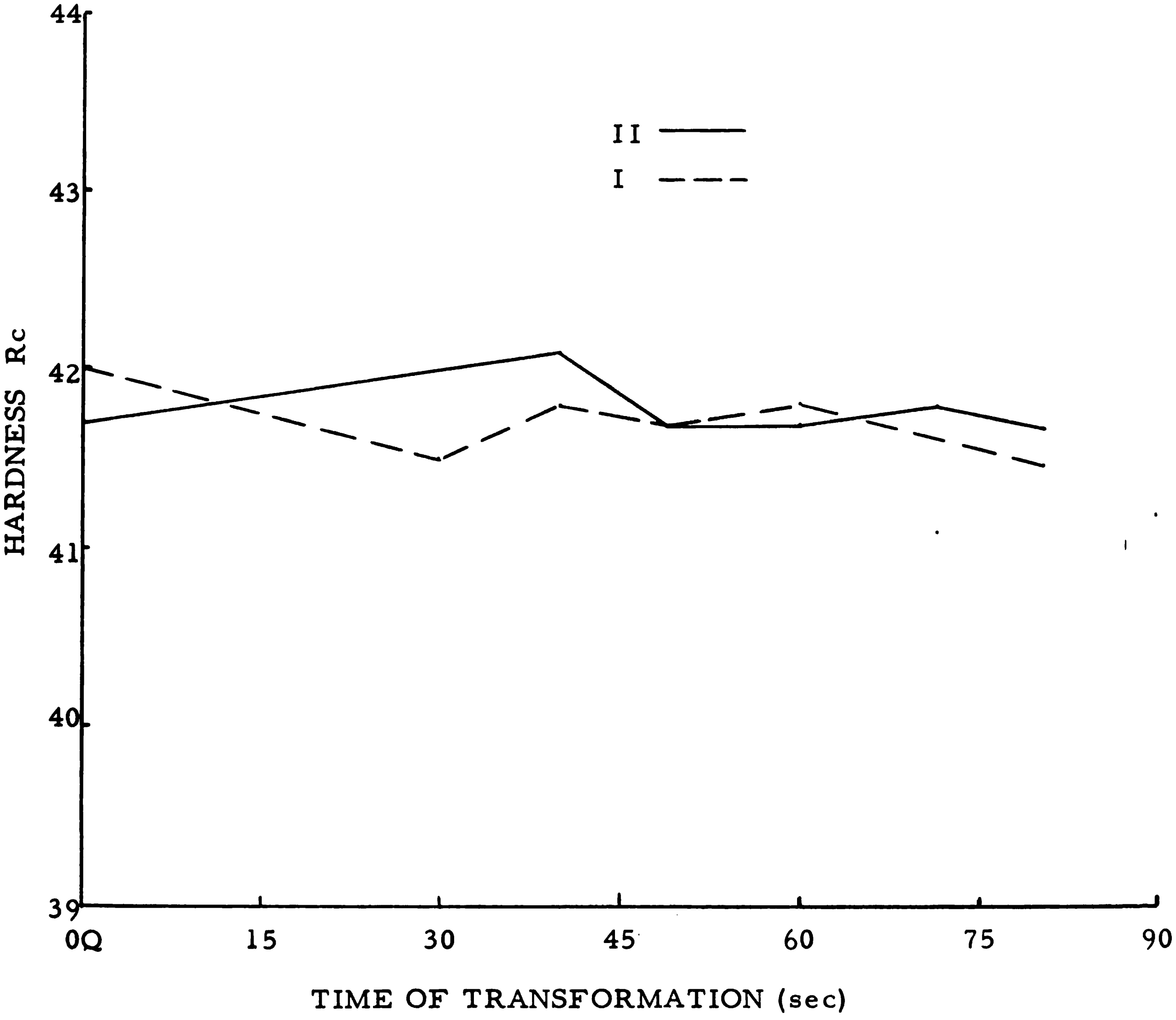

Figure 12 - Hardness range of specimens after tempering.

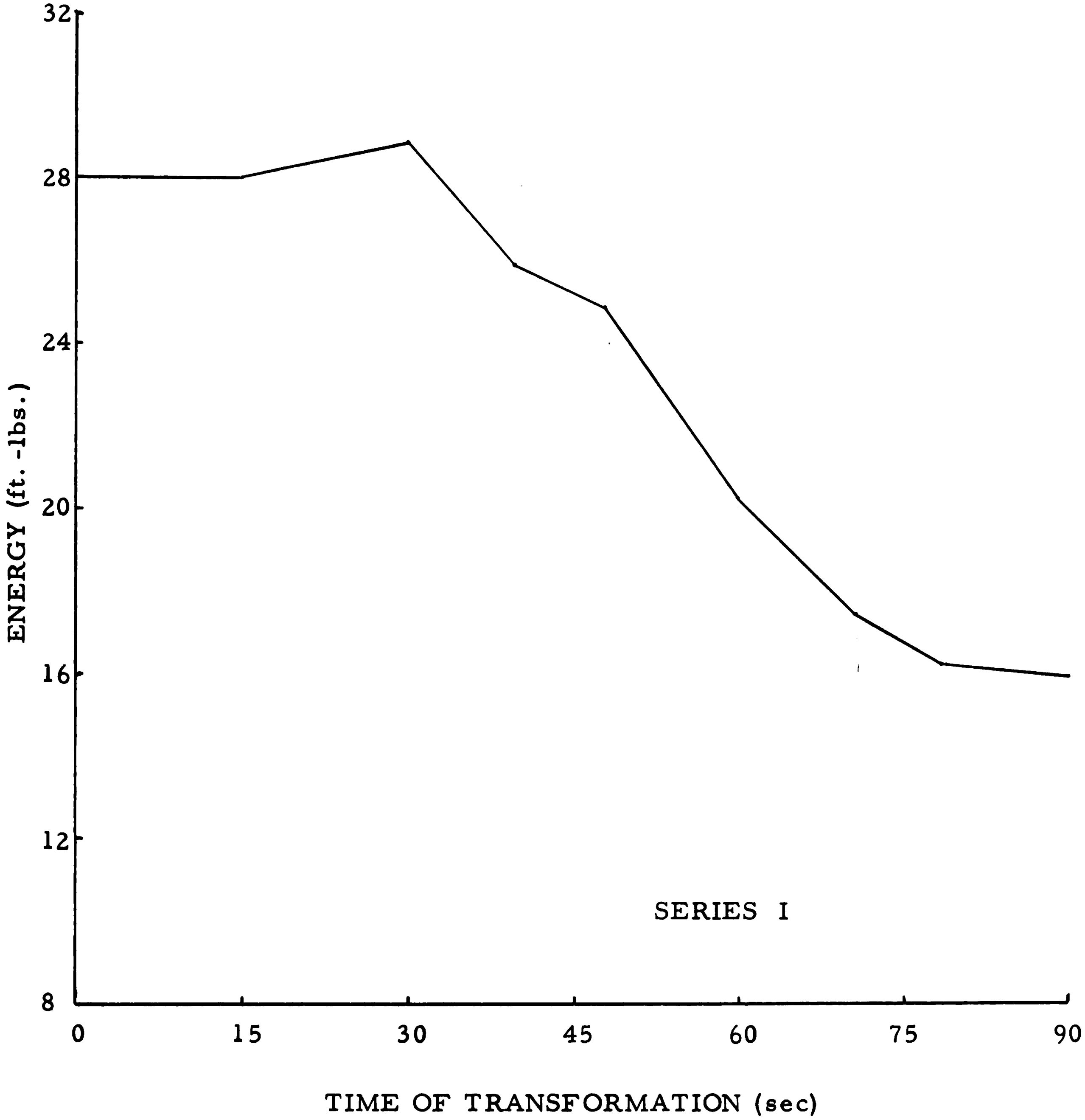

Figure 13 - Room temperature impact strengths vs. time of transformation.

of the amount of bainite produced, hardness readings were taken on samples representing each time of transformation. These values are plotted in Figure 14.

Untempered specimens were examined to allow accurate estimation of the percentage of bainite in the microstructure without the difficulty inherent in differentiating between bainite and tempered martensite. It was further hoped that this technique might give some insight into the morphology of the bainite and thereby explain its detrimental effect. Due to the high hardenability of SAE 4340, it was assumed that the final microstructure of the tempered specimen would consist of tempered martensite and primary bainite only and would not involve secondary bainite as in the case of bainitic hardening of steel. Microscopic examination of the specimens supported this assumption. It was further assumed that since specimens for any one series of experiments were all taken from the same bar, the kinetics of the bainitic transformation would not vary appreciably from specimen to specimen and therefore, specimens which were subjected to identical heat treatments could be assumed to contain identical quanitities and distributions of microconstituents.

At this stage of the investigation, it appeared certain that there was a deleterious effect on toughness caused by the presence of small amounts of bainite in a matrix of tempered martensite. It also appeared that at room temperature the critical amount was that formed in about one minute at 825°F. These results indicated the range of decomposition product which would have a marked effect on the toughness throughout the temperature range, i.e. would alter the shape of the transition curve. The next step was to prepare a series of specimens consistent with the established heat treating procedures and conduct impact tests

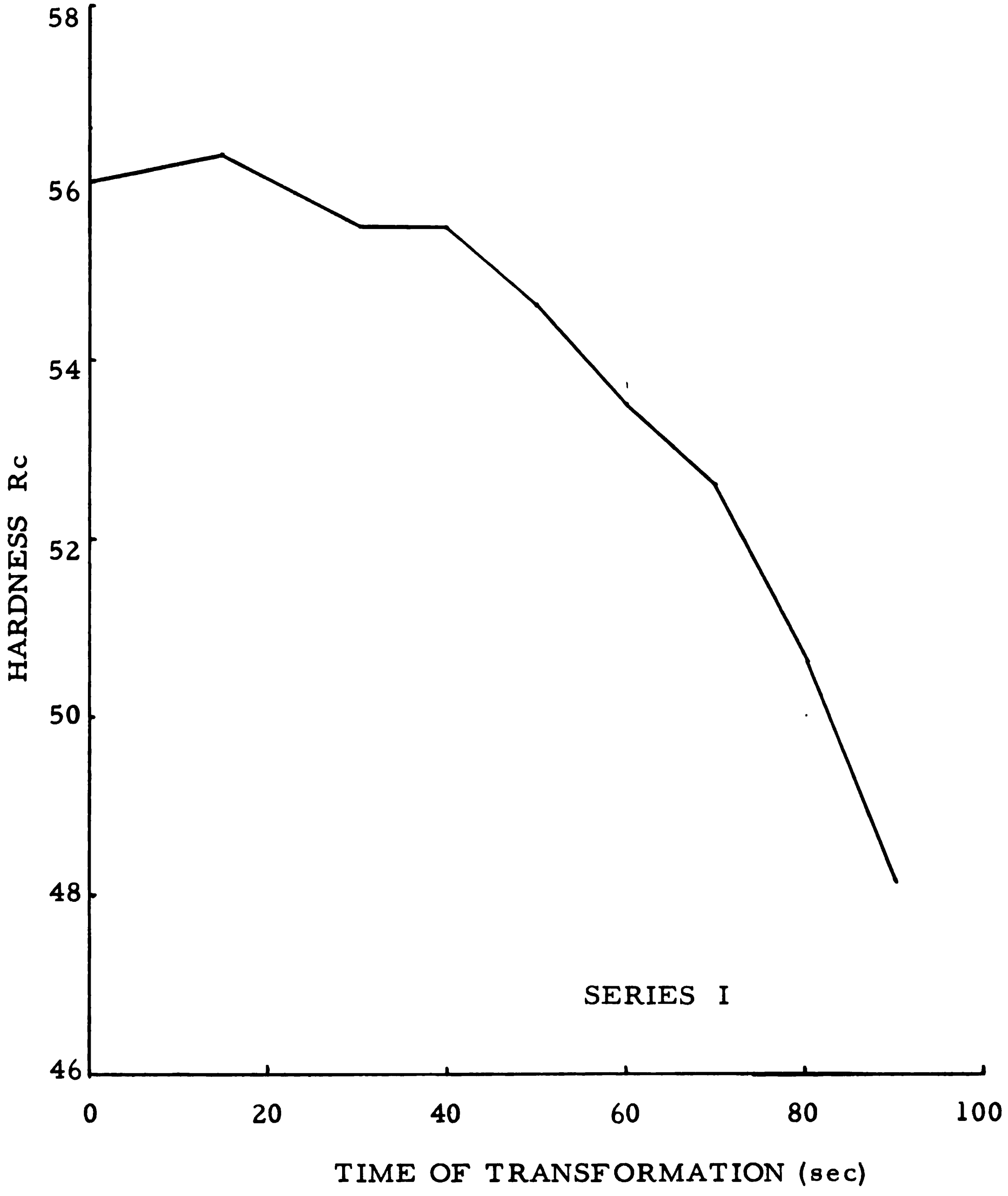

Figure 14 - As-quenched hardness.

at various temperatures so as to determine the transition curves for these various admixtures of microconstituents. Specimens 57-92 were prepared and encompassed the range from direct oil quench to isothermal transformation at 825°F for 80 seconds. Specific heat treatment was as follows:

Specimen	Austenitize	Iso. Trans.	Temper (1 hr)
57-62	30 min at 1550°F	Oil quench	900°F
63-68	"	40 sec. 825°F	875°F
69-74	"	50 sec. "	"
75-80	"	60 sec. "	850°F
81-86	"	70 sec. "	800°F
87-92	"	80 sec. "	800°F

SERIES I I

All specimens were water quenched from the tempering treatment. Hardnesses ranged from Rc 41.5 to Rc 42.3 which was deemed a sufficiently narrow range to preclude hardness from being a significant variable.

These specimens were then broken and the impact energies recorded. Test temperatures were again chosen to be representative of the normal range to be encountered in service. Temperatures of -108, 32, 70, 150, 250, and 350°F were used. Results are shown in Figure 15.

There is clearly a difference between the transition characteristics of those specimens transformed for 60 seconds or less and those which were transformed for 70 seconds or more. Although all curves continue to rise with increasing test temperature, the initial point of breakover into what approximates a plateau region, occurs at approximately

250° F for times of 70 seconds and more whereas this point occurs at about 100° F for those specimens transformed for 60 seconds or less. This is a significant difference. If this breakover point is the definition of transition point, it is clear that the critical amount of bainite has been exceeded in the lower set of curves. If the 15 ft.-lb. criterion is used, it is still obvious that the two sets of curves differ markedly. The lower set of curves crosses the 15 ft.-lb. line at approximately 60° F while the upper set of curves does not reach this low level until somewhat below -100° F.

There is a serious loss of toughness in SAE 4340 steel when too much upper bainite is present in the microstructure. There is also a marked change in the entire transition curve for the steel with the critical amount of upper bainite causing a significant increase in the transition temperature. The reason for this behavior is not known although a theory will be offered in the conclusions drawn from this work.

ANALYSIS OF RESULTS

Results of this investigation support the contention that 100% bainitic steel is superior to a quenched and tempered steel at the same strength level. Even though there was a slight hardness differential between the two series of specimens, the data shown in Figure 11 indicates that the impact values of the two series would be even more at variance with one another if the specimens had been at identical hardnesses. The reason for this superiority of bainite over tempered martensite may be the absence of sub-microscopic cracks in the bainitic steel which could be generated in the quenched and tempered martensitic steel. Early work by Davenport and Bain (6) demonstrated the effect

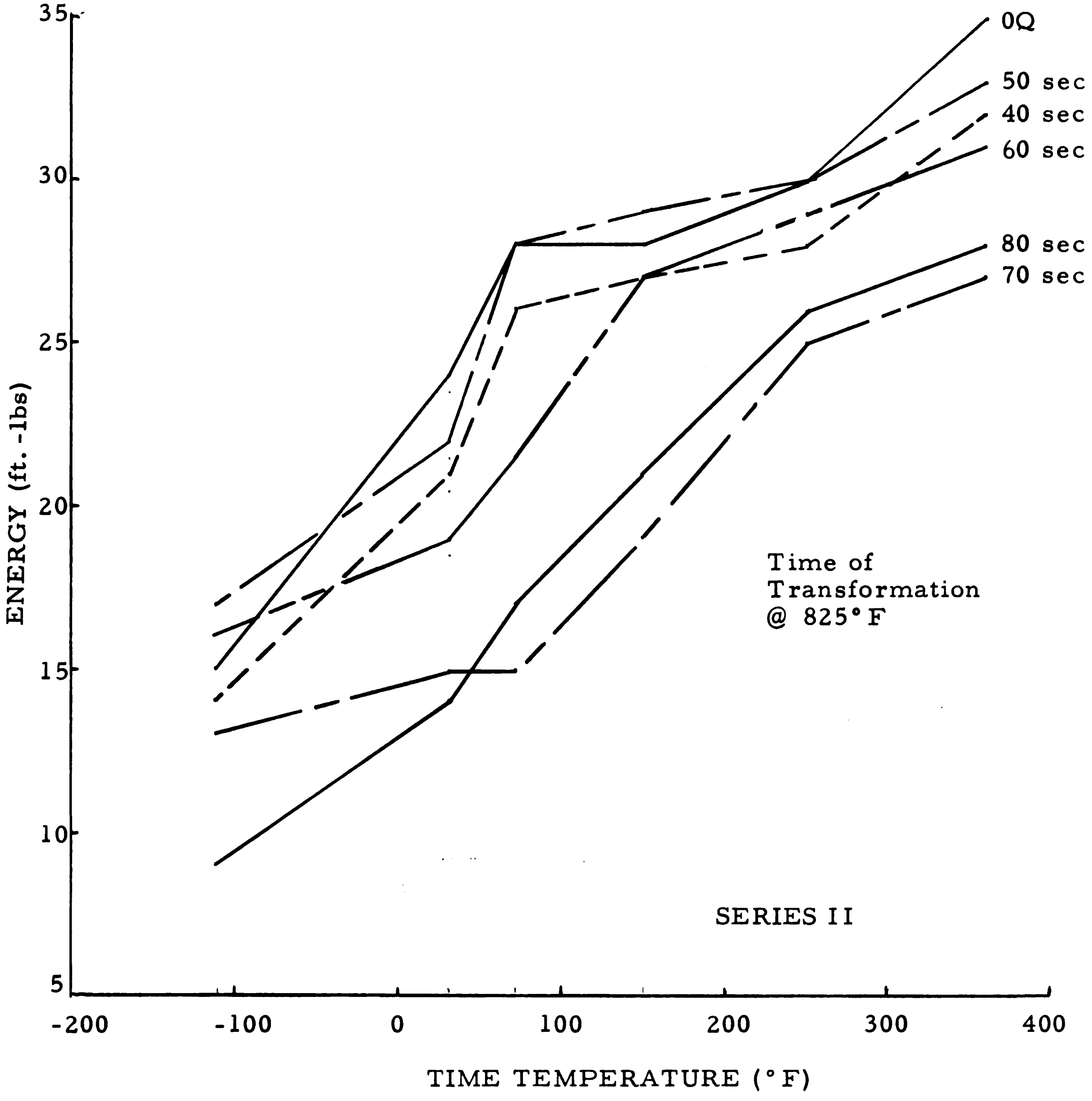

Figure 15 - Transition curves for varying derivatives of bainite and tempered martensite.

of microcracks in a quenched structure. Due to the sudden quenching and the resultant high internal stresses which are generated during the rapid transformation of austenite to martensite, it is conceivable that many small cracks could be formed. These cracks may then serve as internal stress risers and thereby reduce the toughness of the steel. It has been shown that often extremely high tempering temperatures are needed to eliminate these cracks in some steels. When steels are to be used at very high strength levels, it is obvious that the use of high tempering temperatures is impossible. The bainitic structure appears to possess the toughness and strength properties of the martensite but due to the formation of the bainite resulting from isothermal transformation above Ms under nearly equilibrium conditions, there is a net effect of forming a structure of great strength and toughness without the ill effects of microcracks resulting from quenching. It is recognized that this is a very old theory and that it has received little support in recent years. No other firm explanation to account for the superiority of bainite has been offered, however.

There is much discussion in the literature as to the nature of the martensitic transformation in steel. It has often been described as a shear phenomenon involving a complex shear within the crystal structure which results in an almost instantaneous transformation from face centered cubic austenite to martensite which is a body centered tetragonal structure of alpha iron supersatureated with carbon. It is also claimed that the transformation may involve an extremely rapid nucleation and growth process. The principle factor in this mechanism is that iron exists in FCC form at the austenitizing temperature but is BCC at temperatures below the critical. The fact that the carbon atoms are trapped within the lattice due to the rapid structural change which occurs upon quenching and are thereby prevented from diffusing to more favorable lattice sites, accounts for the distortion of the normal low temperature body centered cubic form of iron into a body centered tetragonal form with resulting high hardness. The hardness is a function of the degree of tetragonality and therefore of the carbon content of the steel.

The tempering of steel involves reheating of the specimen to some temperature below the critical and holding for a given period of time. In this work, one hour was used as a standard time of tempering. At elevated temperatures, the lattice loosens to the point where the entrapped carbon atoms can diffuse to more favorable sites, thereby lowering the degree of tetragonality, the lattice strain, and the hardness and strength.

Tempering is generally accepted as involving three distinct stages. The first is the precipitation of epsilon carbide. This hexagonal form of iron carbide is found as the specimen is heated through the range of temperatures from 170-320° F and accounts for a degree of precipitation hardening of the steel. The second stage of tempering occurs in the temperature range 450-540° F and involves the decomposition of any retained austenite and its subsequent transformation to a structure resembling bainite. The third and final stage corresponds to the formation of cementite from the epsilon carbide and occurs in the range of 540-720° F. A fourth stage of tempering is referred to in Hume-Rothery's book, "The Structure of Metals and Alloys." (26)

As already pointed out in this paper, the production of lower bainite at temperatures near 600° F involves the formation of a structure with habit and morphological characteristics which make it strikingly similar to a high carbon martensite. To review briefly, it was shown by Hehemann et al that both high carbon martensite and lower bainite have truly plate-like structures and grow with a $\{225\}_\gamma$ or other high order habit. Also it was pointed out that the production of bainite at such temperatures involved carbide phase transformations similar to those which occur in the first and third stages of tempering. Thus, epsilon carbide, which is formed initially, transforms to cementite during continued holding at temperatures in the vicinity of 600° F. Since the rate of precipitation of epsilon carbide as well as its transformation to cementite is a diffusion controlled process it is highly temperature dependent.

Lower bainite and a tempered high carbon martensite are very similar. Lower bainite appears to have an advantage, however, in that it demonstrates superior toughness throughout the normal service temperature range. It is apparent then, that an austempered structure may be superior in toughness and strength to tempered martensite for many applications. There are two principle limitations to the austempering technique, however. First, there are relatively few steels which can be transformed to 100% bainite at any temperature and none that can be transformed to a fully bainitic structure throughout the temperature range of bainite formation. This is significant since admixtures of bainite and tempered martensite can be deleterious to mechanical properties, especially toughness. A second limitation is the maximum hardness or strength level which can be achieved using this technique. The data in this paper show that an oil quenched SAE 4340 yields an as-quenched hardness of Rc-56 whereas the maximum hardness obtainable by austempering is about Rc-49.

In this age when the demand for high strength steels is great, this limitation in strength may well prevent the use of austempering for some purposes. Although strengths of 200,000 psi should be attainable in SAE 4340, it must be remembered that the probability of brittle failure increases as the strength level is increased. Current work on maraging steels indicates that a major step has been taken in the development of very high strength steels possessing excellent toughness although there is a cost penalty incurred by the use of these new steels.

The second phase of this work also involved the influence of bainite on the impact properties of SAE 4340 steel. In this case, however, it was not a matter of bainite versus tempered martensite, but rather the effect of varying amounts of upper bainite as an admixture in tempered martensite which was of concern. The data clearly indicate that there is a marked detrimental effect on toughness when excessive upper bainite is allowed to form. Although it is contended by others that the embrittling is probably not caused by the presence of bainite as a grain boundary constituent (3) it is accepted that there is a narrow range of decomposition product which, if exceeded, results in serious embrittlement.

It would appear to this author that if the embrittlement of this steel obeyed the law of mixtures as implied in the above mentioned work, the deterioration in toughness would occur gradually and would not exhibit a sudden decrease. It is not within the scope of this paper to determine the origin of upper bainite in steel. It is known, however, that grain boundaries may serve as effective sites of nucleation for other phases and microconstituents which depend upon diffusion for their formation. Thus, it is possible that in upper bainite, where diffusion of carbon atoms is both probable due to the high temperatures involved, and necessary, for growth of the bainite needles, the grain boundaries could serve as diffusion catalysts thereby enhancing one of the conditions necessary for bainite growth, i.e. the lowering of the carbon content in the bainite needle.

Regardless of the acceptance of the grain boundary or mixture concept, it is evident by the microstructures examined that the marked deterioration in toughness and the change in characteristic of the transition curves coincides with the generation of a continuous network of upper bainite. This fact could account for the sudden embrittlement and raising of the transition temperature. Until the bainite becomes a continuous network, the decline in toughness is gradual. Once the network becomes continuous, however, it functions much the same as a grain boundary precipitate in drastically lowering the impact strength of the steel.

Referring to Figures 16 - 23, it is possible to follow the development of the bainitic network in an untempered specimen. Note that the bainite increases from a fine dispersion in Figure 16 to fairly massive areas in Figure 20. It is not until Figure 21, however, that the network of bainite can really be considered continuous. Although this fact may not be readily apparent from this photomicrograph, careful examination of the entire specimen confirmed the assumption of a continuous network. Reference to Figure 15 shows that these two specimens coincide with the series yielding the two characteristic transition curves. That is, specimens with microstructures characterized by Figures 16 - 20 had low transition temperatures and excellent toughness at low temperatures whereas specimens with microstructures represented by Figures 21 - 23

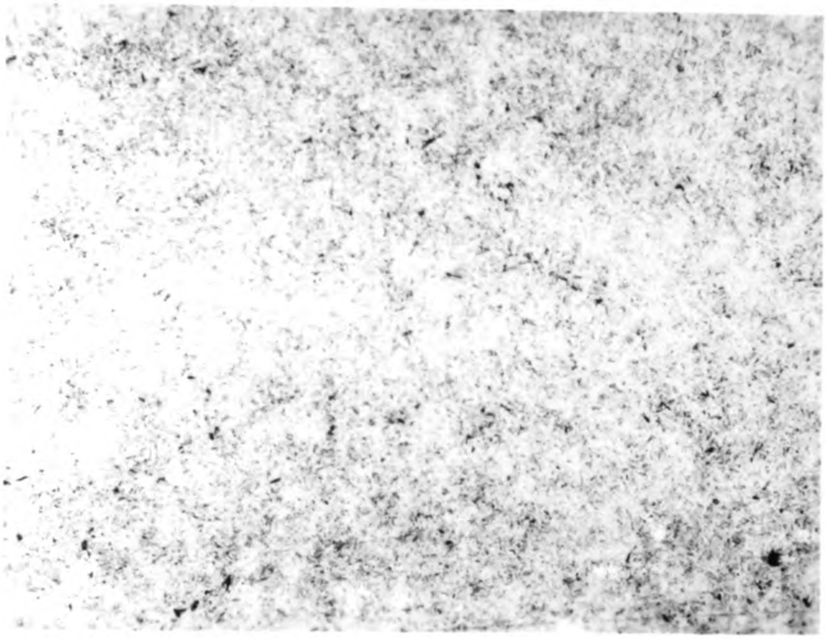

Figure 16 - Microstructure after isothermal transformation for 15 sec. at 825° F. 250x. Nital etch.

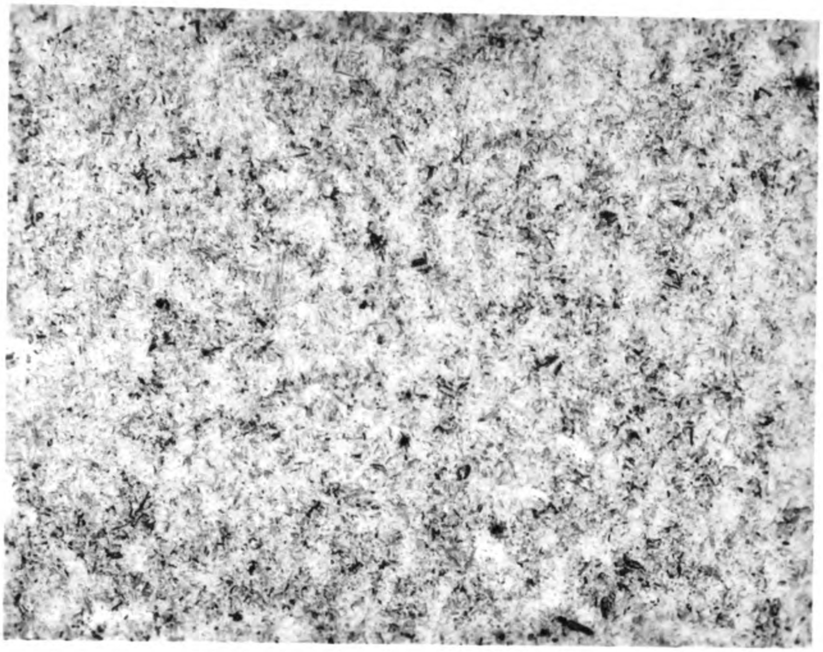

Figure 17 - Microstructure after isothermal transformation for 30 sec. at 825°F. 250x. Nital etch.

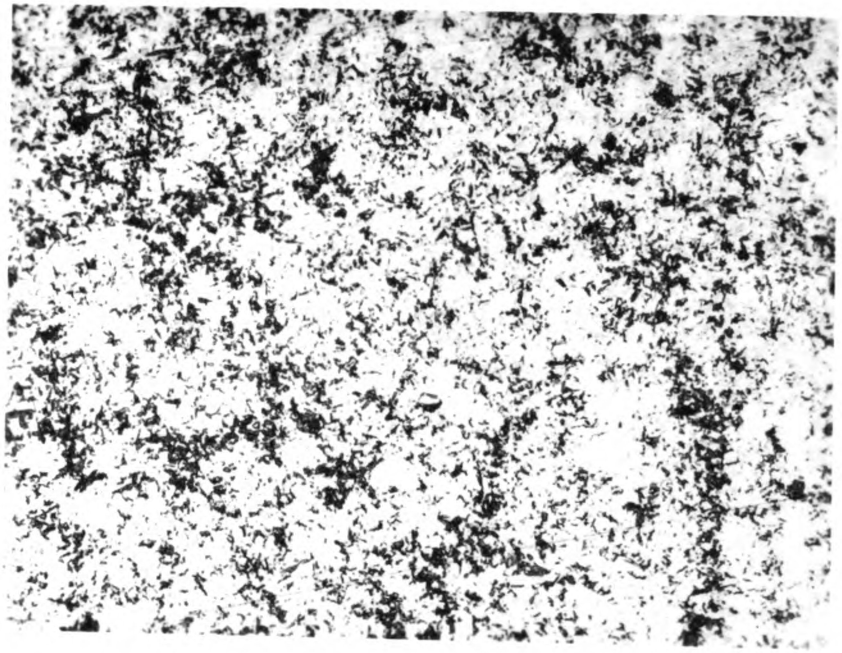

Figure 18 - Microstructure after isothermal transformation for 40 sec. at 825°F. 250x. Nital etch.

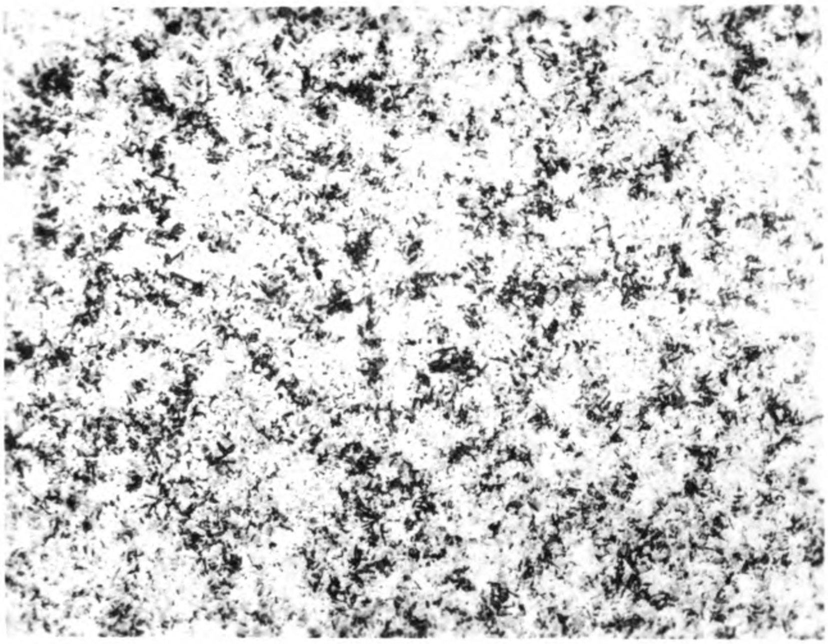

Figure 19 - Microstructure after isothermal transformation for 50 sec. at 825° F. 250x. Nital etch.

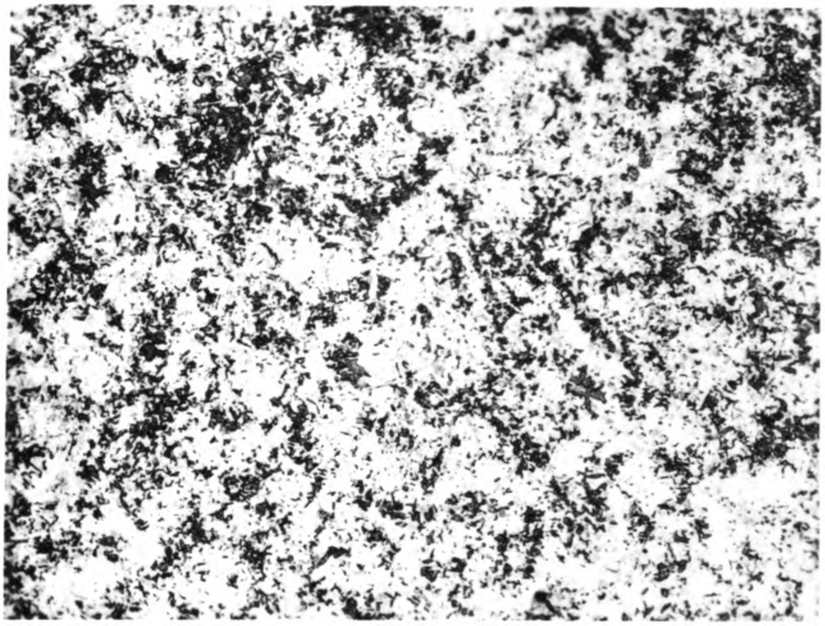

Figure 20 - Microstructure after isothermal transformation for 60 sec. at 825° F. 250x. Nital etch.

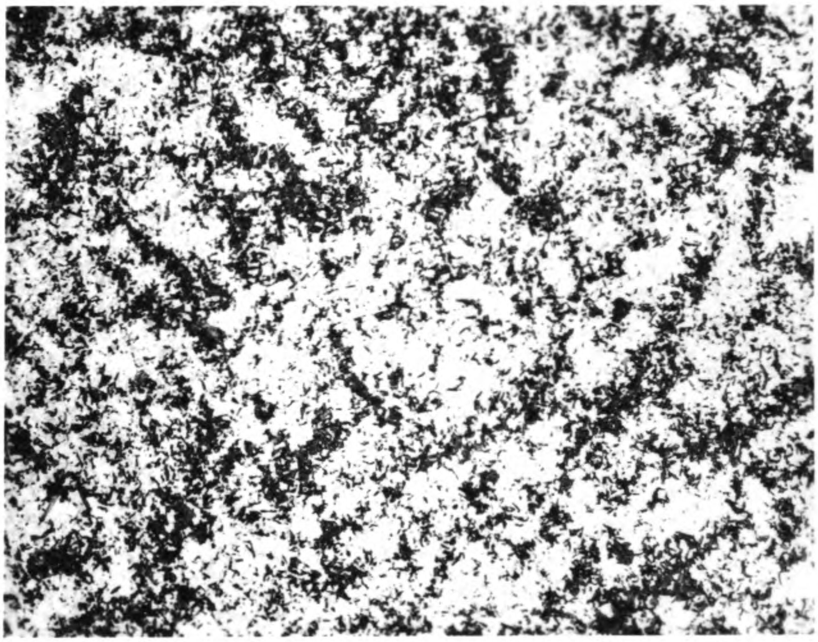

Figure 21 - Microstructure after isothermal transformation for 70 sec. at 825°F. 250x. Nital etch.

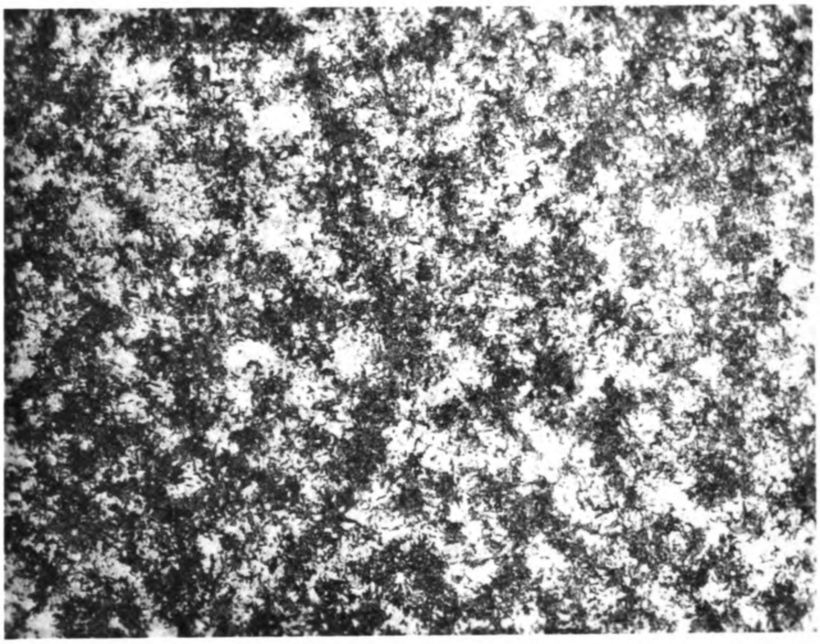

Figure 22 - Microstructure after isothermal transformation for 80 sec. at 825° F. 250x. Nital etch.

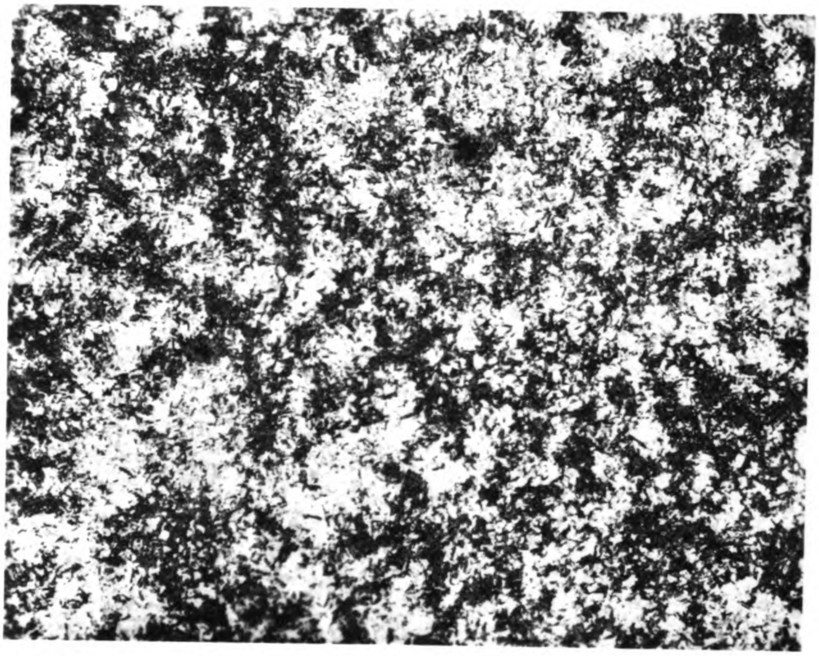

Figure 23 - Microstructure after isothermal transformation for 90 sec. at 825° F. 250x. Nital etch.

had high transition temperatures and yielded impact strengths below the minimum industrial standard of 15 ft. -lbs. at temperatures in the vicinity of room temperature.

Due to the lack of facilities for conducting lineal analysis of these specimens, no accurate estimate of the percentage of bainite present in each of the microstructures was made. It was considered significant that the critical amount of high temperature transformation product was specified by a range of transformation time restricted to a ten second interval.

Based upon limited investigation, there is no reason to believe that lower bainite has the same effect as upper bainite in suddenly reducing toughness, although there is some impairment of this property as the percentage of bainite is increased. It would be of interest to determine why a mixture of tempered martensite, which has good toughness properties, and lower bainite, which has better toughness properties, should yield inferior impact strengths.

CONCLUSIONS

1 - A 100% lower bainite structure possesses higher toughness than a quenched structure tempered to the same strength level. This superiority of bainite exhibited itself throughout the temperature range of the test.

2 - There is a narrow range of upper bainite which, if existent in a matrix of tempered martensite, causes a significant change in the transition curve for SAE 4340 steel. The net effect is to lower the impact values throughout the temperature range and to raise the transition temperature. This effect appears to coincide with the production

of a continuous bainite network. The amount of upper bainite is considerably less than 50%. This suggests that the effect is not dependent upon the law of mixtures but is rather a function of the distribution of the microconstituent.

3 - Although there was a point where impact energies ceased to rise abruptly with temperature, there was no distinct super-transition plateau observed within the temperature range used in this investigation. Impact values continued to rise with increased temperature after the break-over point had been reached.

RECOMMENDATIONS

1 - It is recommended that similar investigations be carried out at other temperatures in the bainite range to determine if a similar embrittlement occurs.

2 - It would be of interest to see if a correlation exists between the true stress-strain characteristics and the impact properties of this steel with regard to the presence of small amounts of upper bainite. Is there any relationship between the yield point phenomenon and the toughness of steels containing both bainite and tempered martensite?

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