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EFFECTS OF TEMPERING ABOVE THE LOWER CRITICAL TEMPERATURE A_{c1} ON THE PROPERTIES OF AN HY-80 STEEL

by

Abner R. Willner

and

Marcel L. Salive

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STRUCTURAL MECHANICS LABORATORY
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Report 2140

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TABLE OF CONTENTS

LIST OF FIGURES

LIST OF TABLES

ABSTRACT

The effects of tempering above the critical transformation temperature on the microstructure, notch-toughness and mechanical properties of HY-80 steel was investigated. The formation of alpha ferrite and prime martensite after tempering above the lower critical temperature and quenching will result in mechanical properties that fall below the requirements of the HY-80 specification. Retempering to achieve minimum yield strengths may result in nonuniform yield strength distribution. The possibility of underbead cracking due to alloy segregation is also discussed.

ADMINISTRATIVE INFORMATION

This work was initiated under Bureau of Ships sponsorship (BuSHIPS letters ALL/NS-011.083 (343) serial 343-211 of 26 May 1959, and R-7-0101 serial 634B-430 of 26 July 1960) and was completed under the David Taylor Model Basin Fundamental Research Program as Problem 735-184, Task 0401, Fundamental Research Project S-R001 01 01.

INTRODUCTION

This is the sixth in a series of preliminary reports[1, 3–6] on a program established to obtain information concerning the effects of nonmartensitic products, chemical composition, and impurities on the notch toughness and weldability of high-strength steels.

A previous report[1] on mechanical properties of commercially produced HY-80 steel indicated that plates which theoretically quench out to 100-percent martensite require higher tempering temperatures than those HY-80 steels containing a small percentage of nonmartinsitic products. In commercial practice, these steels had to be tempered close to the theoretical lower critical transformation temperature A_{e1} to stay below the upper limit of the tensile yield strength requirements of the HY-80 specification MIL-S-16216.[2] The HY-80 steels that were tempered close to, or slightly above, the theoretical A_{e1}* temperature did meet the mechanical property and notch toughness requirements of MIL-S-16216.

The question arose as to what effect tempering HY-80 steel above the actual critical transformation temperature A_{c1}* obtained during heating would have on:

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[1]References are listed on page 23.

*The A_{e1} is the theoretical equilibrium temperature at which carbide begins to transform to austenite upon heating, or the temperature at which the transformation of austenite to ferrite is completed upon cooling; whereas, the A_{c1} is the actual temperature at which austenite begins to form upon heating. The A_{c1} temperature is higher than the A_{e1} temperature because of nonequilibrium conditions existing during heating.

1. Stress-strain curve.
2. Compressive and tensile yield strengths.
3. Nil-ductility-transition (NDT) temperature.
4. Longitudinal Charpy V-notch energy at −120 F.
5. Maximum transverse Charpy V-notch shelf energy.

This report presents the studies made on a fully quenched HY-80 steel tempered both below and above the lower critical transformation temperature A_{c1}, which is approximately 1330 F. Limited studies were made to investigate the effects of retempering at 1250 F on those specimens which were initially tempered above the A_{c1}. These data will be utilized for evaluating HY-80 steels used in model construction. At times it is necessary for the Model Basin to retemper HY-80 steel plates to an 80,000-psi compressive yield strength level, and this necessitates tempering below the A_{c1} temperature for extensive periods. If a double temper (one above the A_{c1} and then a second temper below the A_{c1}) could be performed without adversely affecting the mechanical and notch-toughness properties, considerable time would be saved.

MATERIALS, SPECIMENS, AND PROCEDURES

The HY-80 steel test specimens and test procedures are the same as used by the Model Basin in other related investigations.[3-7] That is, mechanical property test specimens were cut from drop-weight specimens which were heat treated in accordance with the steps shown in Figure 1. Austenitizing was performed in a neutral salt bath furnace. After one-half hour at the austenitizing temperature, the drop-weight specimens were quenched in a brine solution and immediately transferred to a mixture of dry ice and acetone at −110 F and held for 1 hour. All tempering was performed in a neutral salt bath at the temperatures depicted in Figure 1 for 1 hour and then water quenched. For immediate reference, Figure 1 lists the chemistry, heat treatment, and tests performed.

ANALYSIS OF TEST RESULTS

In reviewing the data presented herein, it must be remembered that the data obtained are directly related to the transformation products (alpha ferrite and prime or untempered martensite) produced by tempering and quenching between the A_{c1} and the A_{c3}* and to the effects that a subcritical retemper will have on these transformation products.

*The A_{c3} is the temperature at which the transformation from ferrite to austenite is completed during heating.

METALLOGRAPHY

Figure 2a shows that tempering up to 1330 F produces a progressive spheroidization of carbides. At 1360 F, widmanstatten alpha ferrite becomes evident as is shown in Figure 2b. This indicates that the lower critical transformation temperature A_{c1} falls between 1330 and 1360 F. It is interesting to note that using the Lambert and Grange method[8] gives a calculated A_{e1} of 1327 F. The photomicrograph of the metallurgical transformation produced by heating to 1420 F for 1 hour and then water quenching indicates that the upper critical transformation temperature A_{c3} is being approached. Although not shown in Figure 2, the metallographic structure after quenching from 1640 F is all martensite. No tempering temperatures between 1420 and 1640 F were investigated to determine the A_{c3} by metallographic analysis.

It is interesting to note from Figure 2b that the widmanstatten alpha ferrite is prominent for those specimens receiving an initial temper of 1360 F and less evident for the higher tempering temperatures. This is readily related to the classical iron-carbon phase equilibrium diagram which shows decreasing amounts of alpha ferrite in the alpha-gamma region with increasing temperatures. The exact quantity of alpha ferrite present could not be precisely determined because of the severe microstructural banding which occurred when tempering in the A_{c1} to A_{c3} temperature range.

Figure 3 depicts the coalescense of carbides that occurred when specimens which were initially tempered at 1360, 1390, and 1420 F were retempered at 1250 F.

YIELDING CHARACTERISTICS

Figure 4a indicates the effects of alpha ferrite and prime or untempered martensite on the tensile and compressive stress-strain curves. Specimens tempered below the A_{c1} transformation temperature (1270 to 1330 F) have a plateau yielding characteristic, and specimens tempered at 1360 F and above show curvilinear stress-strain curves.

Although the yielding characteristics for both the tensile and compressive stress-strain curves are the same, the compressive yield strengths are slightly higher than the tensile yield strengths obtained from those specimens tempered between 1270 and 1330 F. In contrast, there is a marked increase in the compressive yield strength over tensile strength for those specimens quenched after tempering at or higher than 1360 F. This difference between compressive and tensile yield strength of specimens tempered above the A_{c1} is related to the formation of high residual tensile stresses in the untempered martensite. These residual tensile stresses result in compressive yield strengths that are markedly higher than the tensile yield strengths.

Tempering of prime martensite at 1250 F reduces a large percentage of residual tensile stresses; in addition, it causes a coalescence of the carbides and thereby lowers the strength of the steel. Figure 4b shows that a plateau or discontinuous yield is obtained after retempering those specimens originally tempered above the A_{c1} transformation temperature; this is similar to the effect that various percentages of proeutectoid or isothermal ferrite and tempered martensite have on the tensile yielding characteristics of HY-80 steel which was tempered at 1250 F.[3]

MECHANICAL PROPERTIES

Figure 4 shows the difference in yield strength due to tempering and quenching from above and below the A_{c1}. The mechanical and notch-toughness properties obtained for these various tempering temperatures are given in Table 1. Figure 5 depicts tensile mechanical properties and nil-ductility-transition (NDT) temperatures after tempering and quenching between 1100 and 1640 F.

Figure 5 shows that the initiation of transition of mechanical properties, determined at ambient temperatures, occurs at 1330 F and is completed at 1450 F, indicating that the A_{c1} and A_{c3} temperatures fall above 1330 F and at or below 1450 F, respectively.

In the previous section on yielding characteristics, it was stated that the formation of high residual tensile stresses in the untempered martensite increases compressive yield strength over tensile yield strength when tempering over the A_{c1} and quenching. The difference and percentage difference in compressive and tensile yield strength after tempering above and below the A_{c1} temperature (taken from the data of Table 1) are plotted for a given tempering temperature in Figure 6a and for a given strength level in Figure 6b.

Figure 6 shows that the compressive yield strength of the as-quenched specimens is 12,000 psi, or 9 percent, higher than the tensile yield strength. Tempering up to 600 F increases this difference to 23,000 psi, or 14 percent; the increase is due to a complexity of metallurgical reactions. In tempering, the crystallographic structure changes from the as-quenched tetragonal to the tempered body-centered cubic. Tempering also causes a precipitation of iron, carbon, hydrogen, and nitrogen compounds, giving rise to two effects–first. the formation of metallurgical keys which prevent dislocation movement, and second, the development of a strained lattice structure. The combined effects are to keep the tensile yield strength constant while the compressive yield strength is increased because of residual tensile stresses.

Tempering between 600 and 1150 F gradually decreases the difference between tensile and compressive yield strength as is shown in Figure 6a. Between 1150 and 1250 F, however, there appears to be a leveling off in the difference between the tensile and compressive yield strength; again thi appears to result from the combination of two effects. The first is the precipitation of a complex molybdenum carbide phase which gives rise to a secondary hardening or aging effect, and the second is a softening effect due to the coalescence of carbides during tempering. It is interesting to note that the 9- to 10-percent difference between compressive and tensile yield strength produced by tempering between 1150 and 1250 F falls in the same range as the 1640 F as-quenched fully martensitic structures.

Figure 6 shows that the minimum difference between compressive and tensile yield strength was obtained at the 1300 F temper where the compressive yield strength was 4000 psi, or 5 percent, greater than the tensile yield strength.

The effects of dislocation keys, precipitates, and residual stresses which form prior to yielding are not carried over to affect the ultimate tensile strength of the steel. However, between the ultimate strength and the breaking strength, a new set of metallurgical conditions prevail and affect the true fracture stress and work-hardening characteristics of HY-80 steel. The effects of these various metallurgical factors on the inelastic region of the stress-strain curve will be the subject of another report.

NOTCH TOUGHNESS

The Charpy V-notch toughness requirements for HY-80 steels (MIL-S-16216) were developed by empirically correlating longitudinal Charpy V-notch energy with NDT data. In the last few years, the concept of low shear energy tearing has been developed. The resistance of a steel to tearing in the presence of a notch under yielding conditions has been related to the upper Charpy V-notch energy absorption shelf. This in turn has been related to its drop-weight energy absorption and its resistance to the explosion tear test.[9] Therefore, the effects of austenitizing above and below the A_{c1} on the brittle behavior have been evaluated by NDT drop-weight specimens and resistance to shear tearing by Charpy V-notch specimens.

NDT Evaluation

A previous Model Basin report[3] showed that various percentages of proeutectoid ferrite in fine-grained HY-80 steel either decreased the NDT temperature or had very little effect, depending on the amount of proeutectoid ferrite present. This effect on the NDT temperature was attributed to the presence of the lower strength, higher toughness ferrite which acted as a notch arrestor.

In the present investigation, the presence of alpha ferrite in a matrix of untempered martensite did not appear to detrimentally affect the NDT temperature; see Figure 5. NDT temperatures of the steels were at or below −120 F regardless of strength or tempering temperature.

The transition indicated by NDT temperatures in Table 1 and Figures 5 and 7 fell between the tempering temperatures of 1250 and 1450 F. The occurrence of the minimum NDT transition at the 1250 F tempering temperature is probably due to the coalescence of carbides, and the increase in NDT temperature above 1250 F and below the A_{c1} temperature is probably due to the precipitation of a complex chromium-molybdenum-carbide as is evident in the microstructures shown in Figure 2a. Above the A_{c1}, the increase in NDT is due to increasing amounts of prime or untempered martensite

Charpy V-Notch Properties

The previously cited TMB report[3] stated that for those specimens treated to contain given percentages of proeutectoid ferrite, there was no simple correlation between transverse Charpy V-notch properties obtained at the NDT temperature and strength level. By combining all the transverse Charpy energies obtained at the NDT temperature for all specimens containing ferrite, it can be shown that the transverse Charpy V-notch energy at the NDT will be 15 ft-lb within one standard deviation of 2.5 ft-lb or within two standard deviations of 5 ft-lb.

Figure 7 shows that the specimens tempered between A_{c1} and A_{c3} (and thereby containing various percentages of alpha ferrite) had a transverse Charpy V-notch energy of 15 to 19 ft-lb at the NDT temperature. Figure 7 shows that specimens quenched to contain 100 percent martensite and tempered below the A_{c1} had a transverse Charpy V-notch energy of 25 to 38 ft-lb at the NDT temperature. It is interesting to note that tempering above 1270 F and below 1360 F will produce from 45 to 75 percent fibrous fracture appearance in transverse Charpy specimens tested at the NDT temperature.

If the transverse Charpy V-notch lateral expansion is to be considered as a criterion, then two different levels are again obtained, one for those specimens containing alpha ferrite (6 to 8 mils) and one for the all-tempered martensitic structure (12 to 30 mils).

Table 1 shows that the longitudinal Charpy V-notch energy, at −120 F, of the specimens tempered between 1330 and 1390 F fell below 30 ft-lb. However, the specimens tempered at 1420 and 1640 F had a longitudinal Charpy V-notch energy absorption above 30 ft-lb at −120 F; it is interesting to note that these specimens showed approximately 30 percent fibrous fracture or about 70 percent grain.

Figures 8 and 9 indicate that tempering HY-80 from 1100 to 1330 F produced transverse Charpy energy shelves greater than 40 ft-lb at +32 F. In contrast, tempering from 1360 to 1640 F resulted in transverse Charpy maximum energies of 25 ft-lb or less. It should be noted from Table 1 that Charpy specimens tested at +32 F had 100 percent fibrous fracture appearance regardless of the energy absorption or strength level. Table 2 compares the properties of specimens tempered between the A_{c1} and the A_{c3} temperatures to the properties of those specimens fully quenched and then tempered to the same yield strength or ultimate strength level. This table shows that it is better to compare mechanical and notch-toughness properties on the basis of a given ultimate strength rather than on the basis of a given yield strength. Figure 8 shows that when correlating transverse Charpy V-notch energy at +32 F as a function of ultimate tensile strength, the relationship is about linear.

The longitudinal Charpy V-notch energy at -120 F and the transverse Charpy energy at +32 F are compared in Figure 9 for various tempering temperatures. The longitudinal Charpy properties at -120 F were sensitive to the effects of tempering at 1330 F, whereas the transverse properties at +32 F were not affected until tempering at 1360 F. The shift in longitudinal Charpy data of Figure 9 correlates very well with NDT temperature data shown in Figure 5, that is, the high temperature (1250-1360 F) complex chrome-molybdenum precipitates had a tendency to shift the notch-toughness transition to higher temperatures. The lowering of the maximum energy absorption level at +32 F can be attributed to the formation of untempered prime martensite at temperatures above the A_{c1}.

EFFECTS OF RETEMPERING

One objective of this study was to evaluate the effects of retempering HY-80 steel which was initially tempered and quenched between the A_{c1} and A_{c3} temperatures.

Because of the limited material available, this retempering study was conducted at a single temperature, 1250 F. The broken drop-weight specimens from the previous study were retempered, and then mechanical property and transverse Charpy V-notch specimens were made; this precluded any NDT or longitudinal Charpy V-notch tests.

In the previous section on yielding characteristics, Figure 4 showed that the stress-strain curves became plateau or discontinuous after retempering. Table 3 indicates that after retempering at 1250 F, the compressive and tensile yield strengths fell below 80 ksi for the specimens initially tempered at 1360 F, whereas the specimens initially tempered at 1390 and 1420 F had yield strengths from 80 to 98 ksi.

Examining the ratio between compressive and tensile yield strengths in Table 3 shows that the compressive yield strength of the specimens initially tempered between the A_{c1} and the A_{c3} were about 26 percent higher than the tensile yield; after retempering at 1250 F, the compressive yield strengths were only 3 to 6.5 percent greater than the tensile yield strength.

After retempering at 1250 F, values for the transverse Charpy V-notch specimens at +32 F were above 50 ft-lb and he specimens had the appearance of 100 percent fibrous fracture. The specimens initially tempered at 1420 F and retempered at 1250 F had better transverse Charpy properties at +32 F than did the specimens quenched from 1640 F and retempered at 1250 F. In fact, examination of Table 1 shows that fully quenched specimens tempered between 1150 and 1330 F had practically the same energy absorption at 32 F as those specimens tempered above the A_{c1} and then retempered at 1250 F. The increase in toughness obtained for those specimens initially tempered at 1420 F and then retempered at 1250 F could be attributed to a fine austenitic grain that is developed at the 1420 F temperature.

DISCUSSION

A previous Model Basin study[3] demonstrated that difficulties could arise in HY-80 steels if proeutectoid ferrite is present in the microstructure. The deleterious effect of the proeutectoid ferrite is even more pronounced if the prior austenitic grain size is ASTM-4 or coarser. A fine-grained steel made from an HY-80 composition which has been heat treated to contain proeutectoid ferrite and tempered martensite has an unpredictable yield strength and also results in an unpredictable shape for the stress-strain curve. More specifically, tempering between 1000 and 1250 F could cause the yield strength to vary as much as 30,000 psi for an HY-80 steel.

This study used the same HY-80 material (TMB plate E103) as had been used in the previously reported work,[3] but tempering temperatures above the A_{c1} were used to obtain various mixtures of alpha ferrite and prime or untempered martensite. The test results show that the 1360 and 1390 F tempers will meet the tensile yield strength of the HY-80 specification but will not meet the −120 F longitudinal Charpy requirements. The Charpy V-notch properties required for plates more than 2 in. thick (30 ft-lb at −120 F) will be met when tempering above 1390 F, but the yield strength will be well above the permissible specification range. Tempering between the A_{c1} and A_{c3} and then quenching will produce ultimate strengths that are 35 to 50 percent greater than the tensile yield strength; the transverse Charpy V-notch maximum energy shelf is 15 to 22 ft-lb lower than the 40 ft-lb minimum required to prevent low energy shear tearing.

When the HY-80 steel which was initially tempered and quenched from between the A_{c1} and A_{c3} temperatures was retempered at 1250 F, the resulting compressive yield strengths ranged from approximately 76,000 to 98,000 psi. The yield strength level obtained upon retempering was dependent upon the initial tempering, i.e., the amount of untempered martensite produced upon quenching. In other words, the specimens initially tempered at 1360 F and quenched contained more alpha ferrite than untempered martensite, whereas tempering at 1420 F produced more untempered martensite than alpha ferrite.

Retempering at 1250 F of structures containing alpha ferrite gave consistent yield strengths and stress-strain curves. It should be noted that the presence of alpha ferrite may have an adverse effect on the yield strength distribution; depending on the percentage of ferrite present and the tempering temperature, erratic yield strength results and stress-strain curves having different yielding characteristics could result.[3] For example, an HY-80 steel with low pearlitic hardenability or a steel with high pearlitic hardenability and unintentionally heat treated to contain ferrite could have yield strengths that fall below and above 80 ksi in the same plate.

Apparently, the presence of alpha or proeutectoid ferrite would not detrimentally affect the brittle behavior of a fine-grained HY-80 steel. However, the transverse Charpy V-notch energy level for predicting the NDT temperature is considerably lower for specimens containing ferrite (17 ft-lb) than for specimens containing 100 percent martensite (30 ft-lb). It should be understood that the cited energy levels are for the steel plate investigated. The effects of ferrite on the mechanical properties of various commercially produced HY-80 steels will be the subject of another report.

In the presence of ferrite, another factor that would have to be considered is the presence of an alloy-enriched martensite. That is, if proeutectoid ferrite is formed because of, either low hardenability or slow cooling rates during quenching, the transformed ferrite will contain only some quantity of those elements with which it forms solid solutions such as manganese and nickel. The remaining untransformed austenite will be enriched with those alloying elements which have limited solubility in ferrite. Similarly, the alpha ferrite that is formed on heating between the A_{c1} and A_{c3} temperatures will reject by diffusion those elements which have greater solubility in austenite. Upon quenching, the enriched austenite will transform into an alloy-enriched or hard martensite. Therefore, the hardenability of the alloy-enriched banded structure would be increased. Quenched and tempered HY-80 steels containing ferrite and an alloy-enriched martensite will probably have a greater propensity to heat-affected-zone underbead cracking than will HY-80 steels not containing ferrite.

As stated in the introduction, commercially produced HY-80 plates which theoretically quench out to 100 percent martensite require higher tempering temperatures to meet the specification's[2] yield strength range.[1] If these fully quenched plates are inadvertently tempered above the A_{c1}, retempering will be required to meet all of the specifications requirements. Although the retempered plates will meet the mechanical and notch toughness requirements of the HY-80 specification,[2] the propensity to underbead cracking may not be relieved because of the presence of the ferrite and the alloy-enriched tempered martensitic areas. However, to ensure that the A_{c1} temperature is not exceeded, a maximum tempering temperature of 1300 F should be included in the HY-80 specification.

The problem of underbead cracking due to alloy segregation cannot be solved by avoiding the formation of proeutectoid ferrite through an increase in pearlitic hardenability. This would increase the overall hardenability of the HY-80 steel, thus increasing the likelihood that underbead cracking would occur and that the A_{c1} temperature could be exceeded in the tempering. Nor can the problem be solved by decreasing the overall hardenability[1, 3] since the presence of bainite in coarse-grained structures reduces notch toughness considerably.[3] That is, in welding, the probability of obtaining nonmartensitic products in the heat-affected zone is increased.

To ensure that the hardenability of a given thickness of HY-80 steel plate is neither too high nor too low, the two chemistry ranges previously specified in MIL-S-16216D (one for thin and one for thick plate) should be made a requirement of the current specification. These two chemistry ranges should be adhered to until studies are made to evaluate the effects of the presence of nonmartensitic products and alloy enriched martensitic areas in commercially produced HY-80 steel plates on the propensity toward underbead cracking.

Heating an HY-80 steel between the A_{c1} and A_{c3}, quenching, and then retempering to obtain an 80-ksi yield strength is not advised for material to be used in structural model evaluation. The presence of alpha or proeutectoid ferrite will increase the possibility of obtaining inconsistent and nonuniform yield strengths and stress-strain curves as well as the possibility of increasing the presence of underbead weld cracking; these inconsistencies will probably influence the results obtained from static and dynamic model tests.

CONCLUSIONS

1. Tempering above the A_{c1} and quenching produces a mixed widmanstatten alpha ferrite and untempered martensitic microstructure.

2. The as-quenched microstructure containing alpha ferrite and untempered martensite produces:

   a. Curvilinear stress-strain curves.

   b. An increase in both tensile and compressive yield strengths.

   c. An increase in the ratio of compressive to tensile yield strength.

   d. A loss in longitudinal Charpy V-notch energy at −120 F.

   e. Transverse Charpy V-notch energy shelves are below the energy level required for resistance to shear tearing.

   f. Increased NDT temperatures but not to a level where it would be considered critical.

   g. A decrease in transverse Charpy V-notch energy level at the NDT temperature.

3. A better correlation between notch toughness and mechanical properties is obtained when the comparison is based on ultimate tensile strength rather than on tensile yield strength.

4. Retempering of plates with microstructures containing ferrite and prime martensite can produce HY-80 steels that meet the mechanical properties of the HY-80 specification MIL-S-16216.

5. On the basis of this and a previous study,[3] it is concluded that HY-80 steels containing ferrite may be unsuitable for structural applications because of:

a. Nonuniform yield strength distribution.

b. Nonuniform stress-strain characteristics.

c. A possible tendency toward underbead cracking during welding due to alloy-enriched martensitic areas.

6. HY-80 steels containing any alpha or proeutectoid ferrite should not be used in general structural model construction.

RECOMMENDATIONS

1. An upper tempering limit of 1300 F should be made a requirement of MIL-S-16216.

2. The two chemistry ranges, one for thin and one for thick plates, previously specified in MIL-S-16216D should be made a part of the current specification.

3. A double temper, consisting of tempering above the A_{c1} followed by retempering below A_{c1}, should not be used for reducing the yield strength of HY-80 steel.

HY-80 STEEL												
CHEMICAL COMPOSITION, PERCENT												
C	Mn	P	S	Si	Ni	Cr	Mo	Cu	TOTAL Al	ACID SOL. Al	Ti	V
0.16	0.30	0.009	0.017	0.25	2.78	1.52	0.36	0.058	0.042	0.038	0.001	0.001

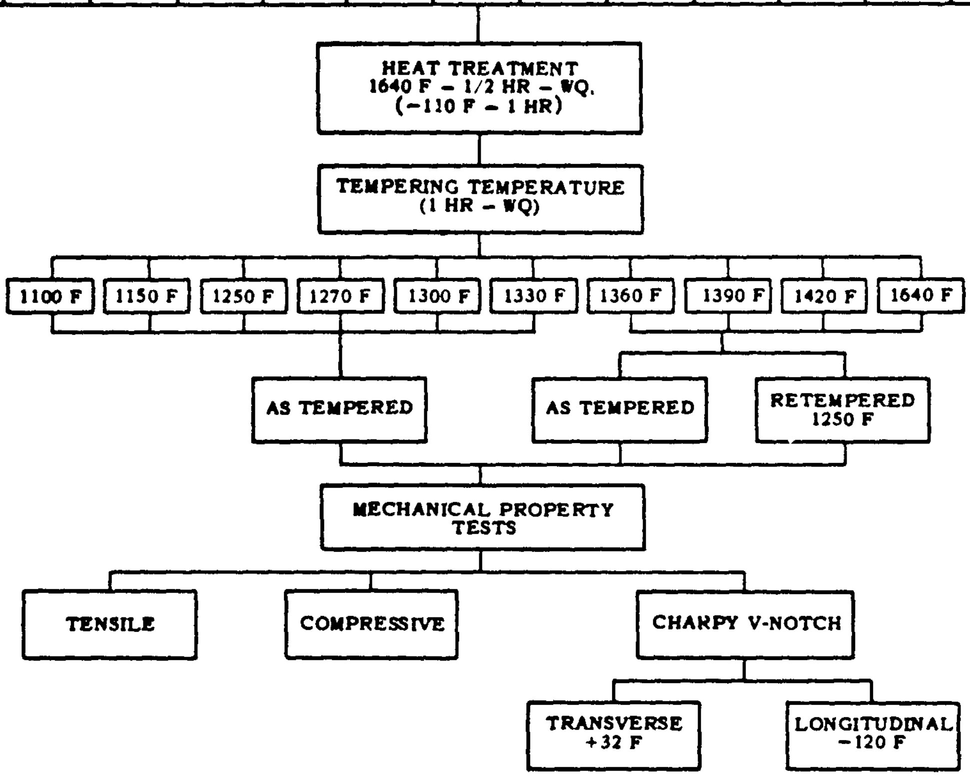
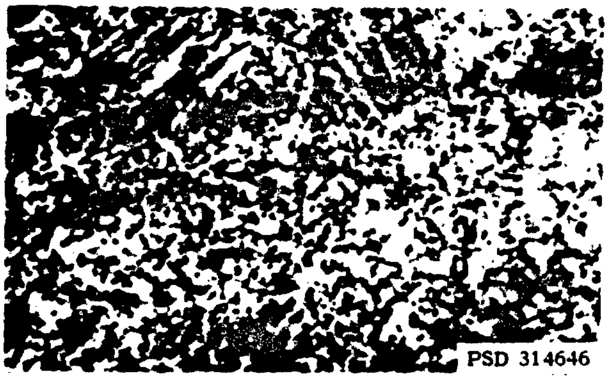
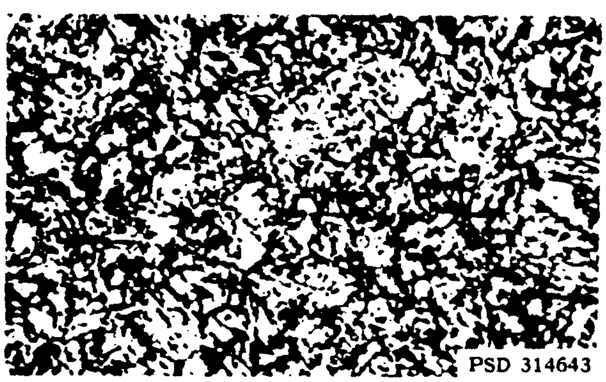


Figure 1 – Chemical Composition of 5/8-Inch-Thick HY-80 Steel Plate and the Investigative Steps Followed in This Study



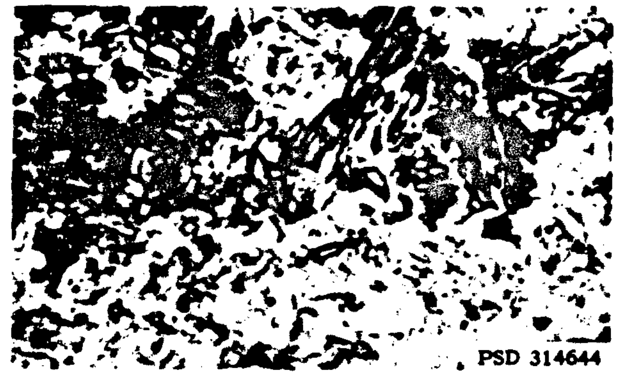
1330 F

1300 F

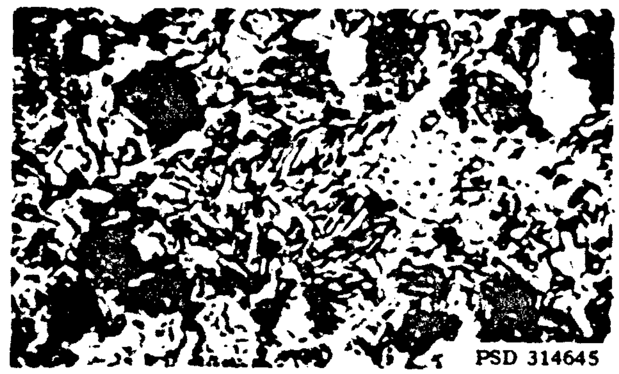


1270 F

Figure 2a – Temperatures below the A_{c1} Temperature



1420 F

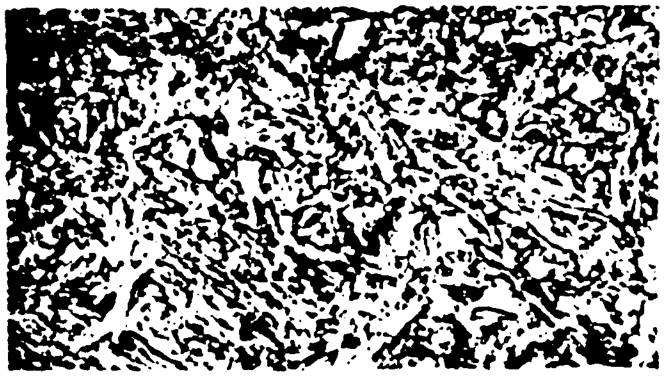


1390 F

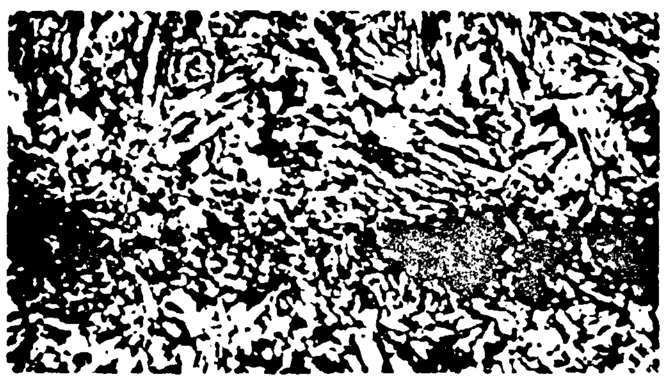
1360 F

Figure 2b – Temperatures above the A_{c1} Temperature

Figure 2 – Microstructure (100X) of Specimens Tempered below and above the Lower Critical Transformation Temperature A_{c1}



Initial Temper 1420 F 1 Hour
Retemper at 1250 F 1 Hour, W.Q.



Initial Temper 1390 F 1 Hour
Retemper at 1250 F 1 Hour, W.Q.

Initial Temper 1360 F 1 Hour
Retemper at 1250 F 1 Hour, W.Q.

Figure 3 – Microstructure (1000X) of Specimens Retempered a' 1250 F after an Initial Tempering above the Low-. Critical Transformation Temperature

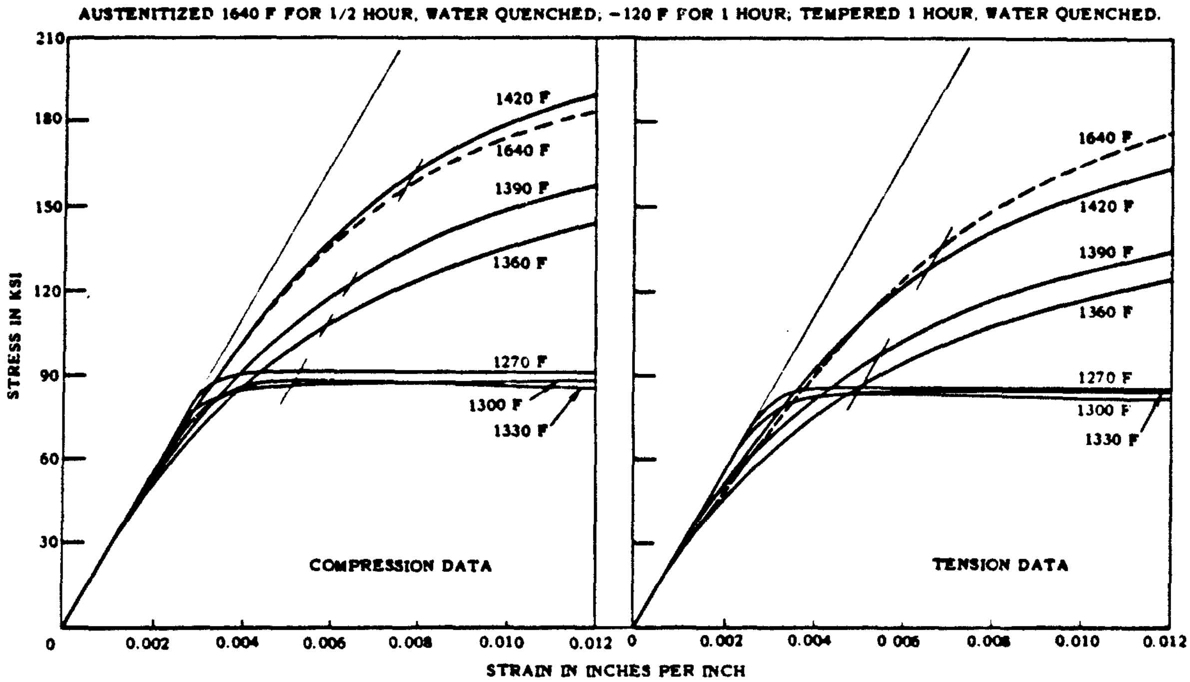


Figure 4a – Stress-Strain Curves Produced by Initial Tempering above and below the Lower Critical Transformation Temperature

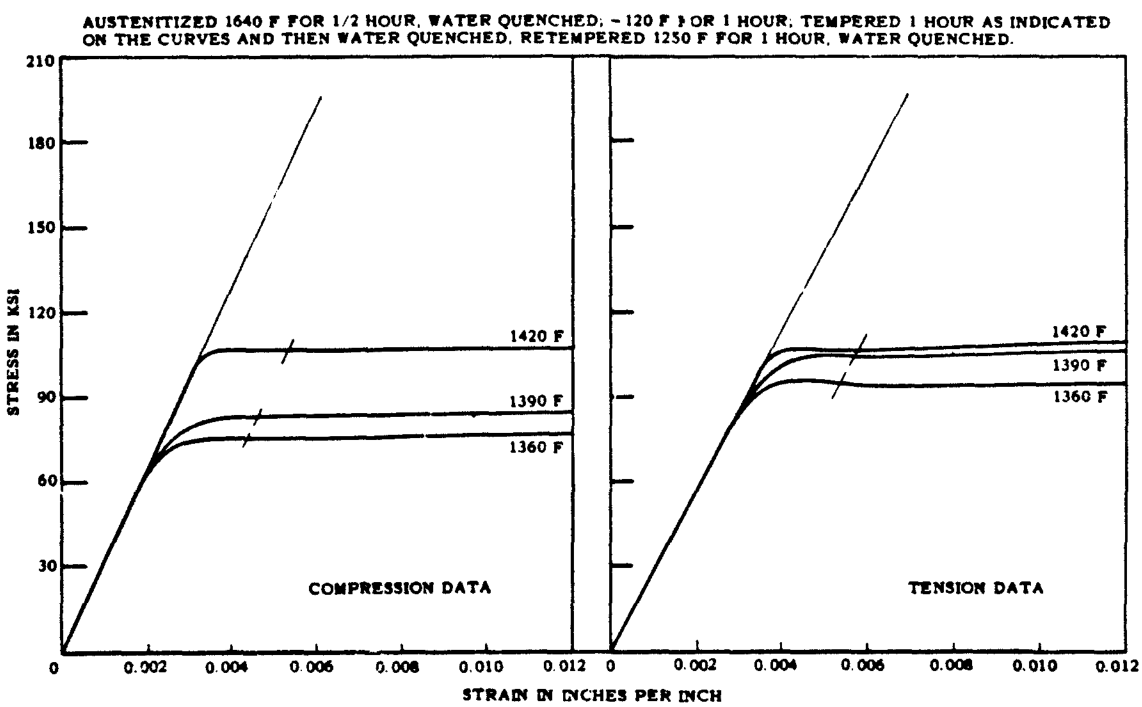


Figure 4b – Stress-Strain Curves of Specimens Retempered at 1250 F after an Initial Tempering above the Lower Critical Temperature

Figure 4 – Tensile and Compressive Stress-Strain Curves Produced by Various Thermal Treatments

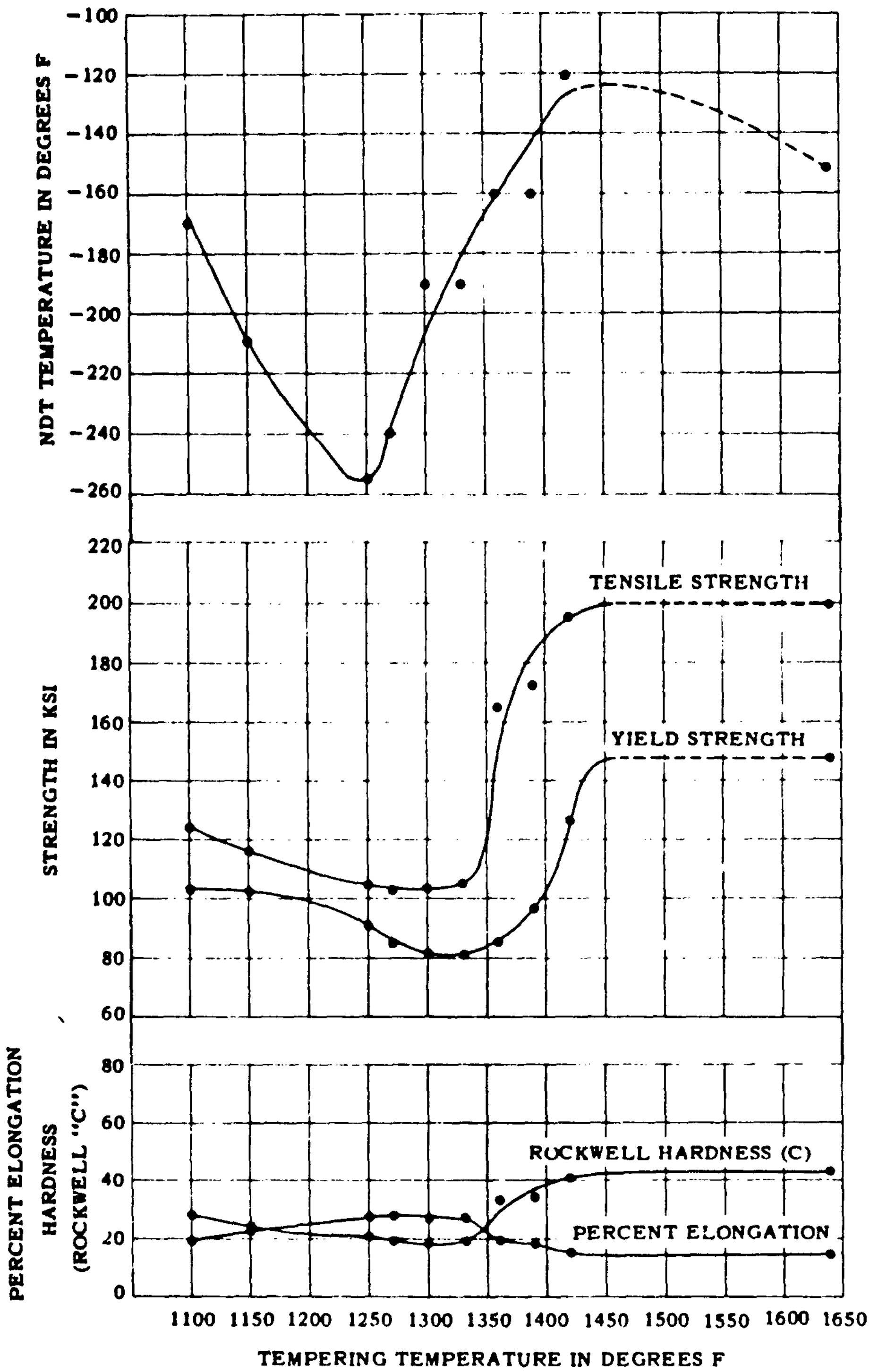


Figure 5 – Effects of Tempering Temperature on the Mechanical Properties and NDT Temperature of an HY-80 Steel

Figure 6 – Effects of Tempering Temperature and Strength on the Relationship between Tensile and Compressive Yield Strength

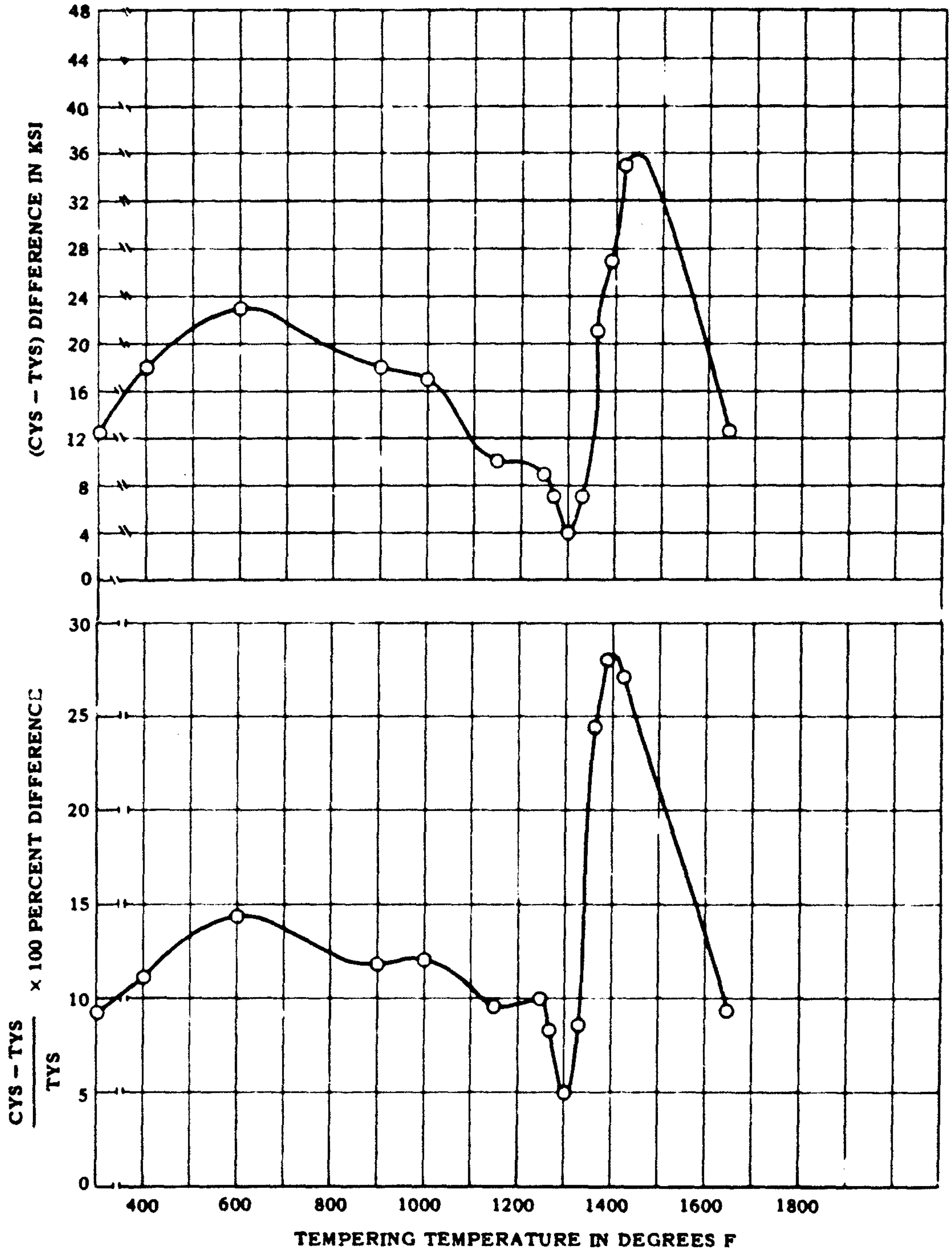


Figure 6a – Effects of Tempering Temperature on the Relationship between Tensile and Compressive Yield Strength

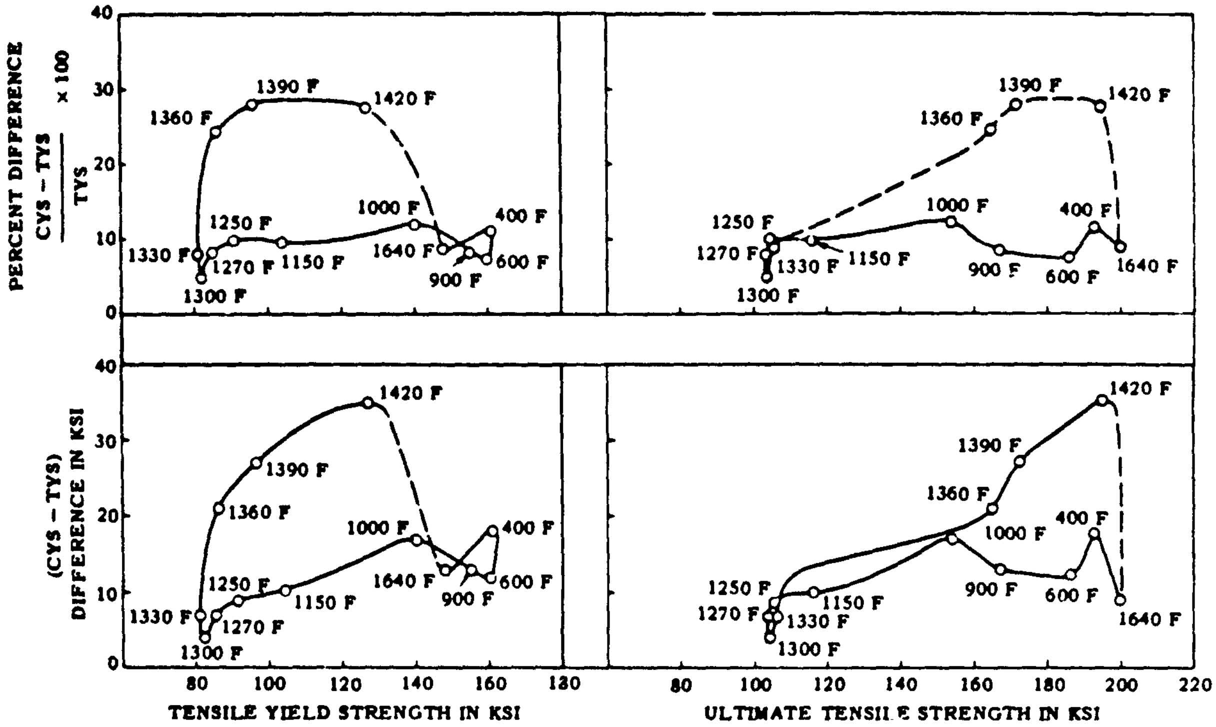


Figure 6b – Effects of Tensile Strength on the Relationship between Tensile and Compressive Yield Strength

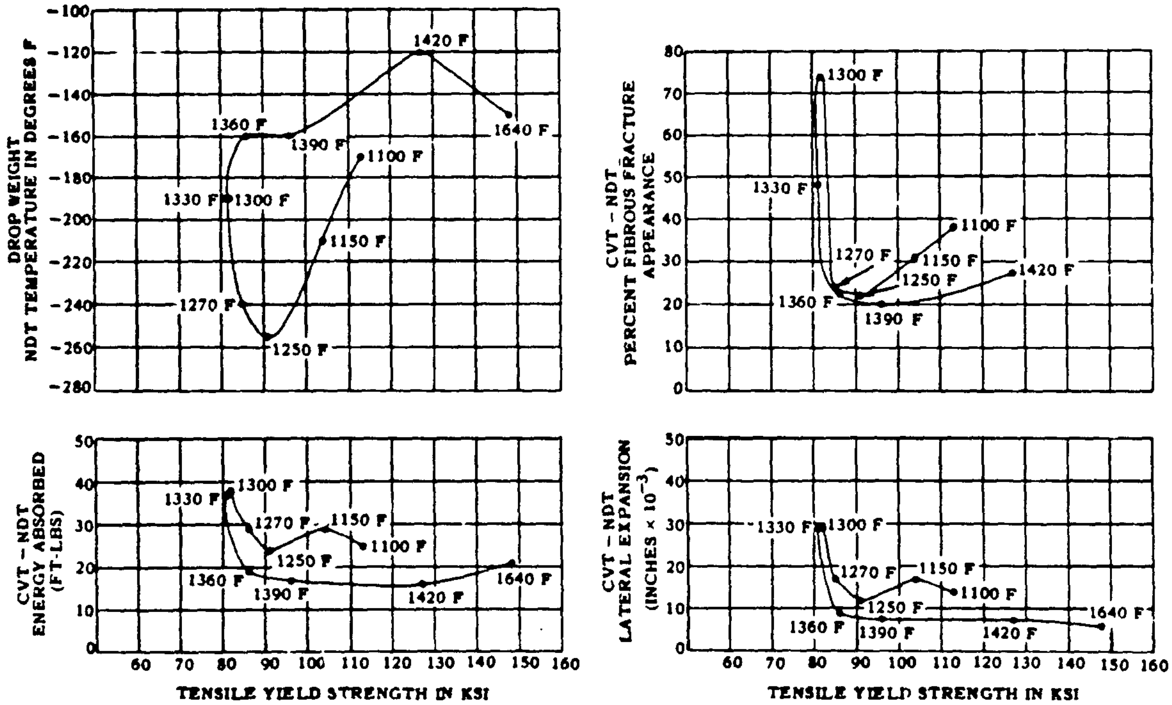


Figure 7 – Effects of Tensile Yield Strength on the NDT Temperature and on the Transverse Charpy V-Notch Properties at the NDT Temperature

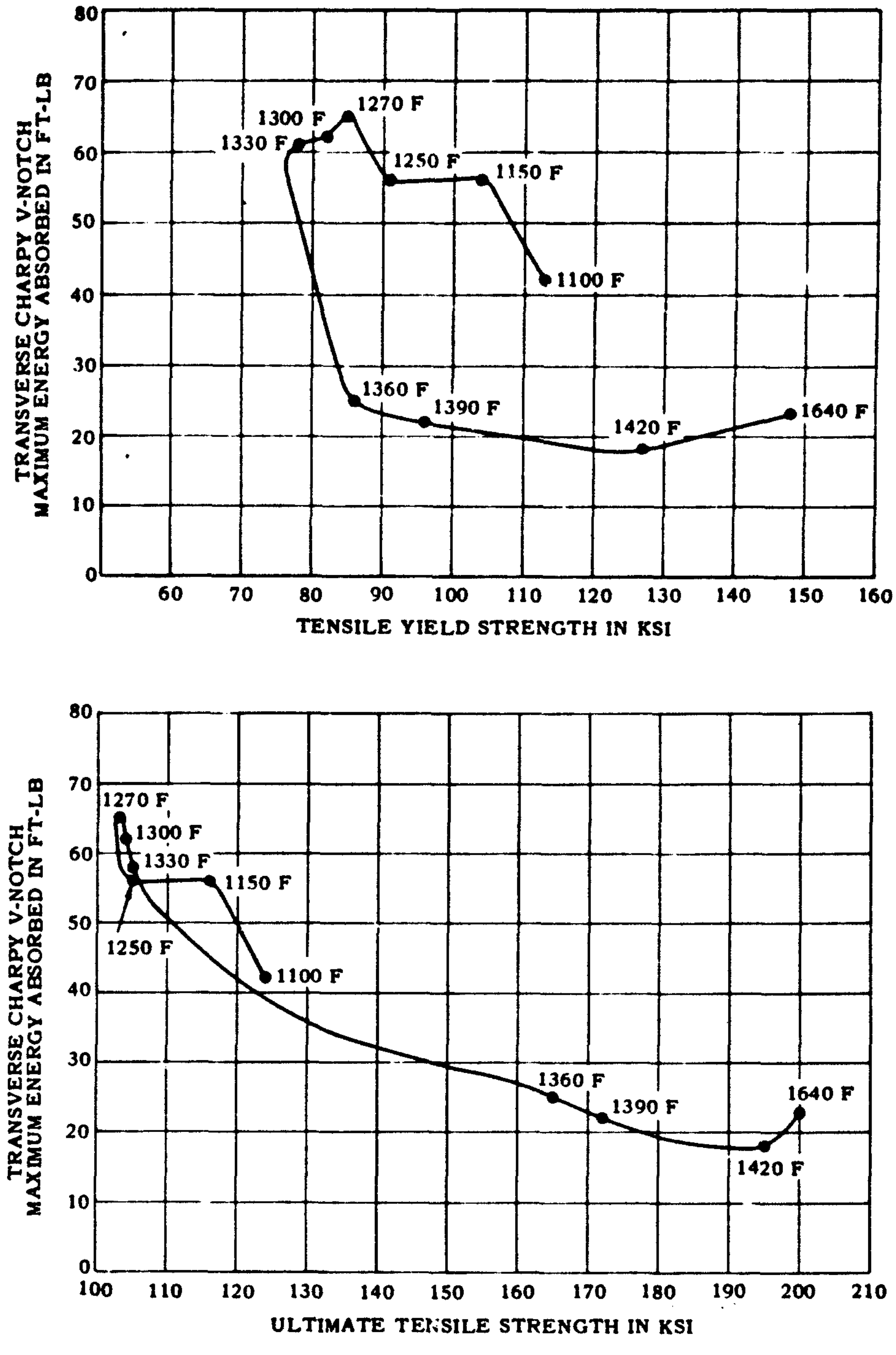


Figure 8 – Effects of Strength Level on the Transverse Charpy V-Notch Maximum Energy Level Measured at 32 F

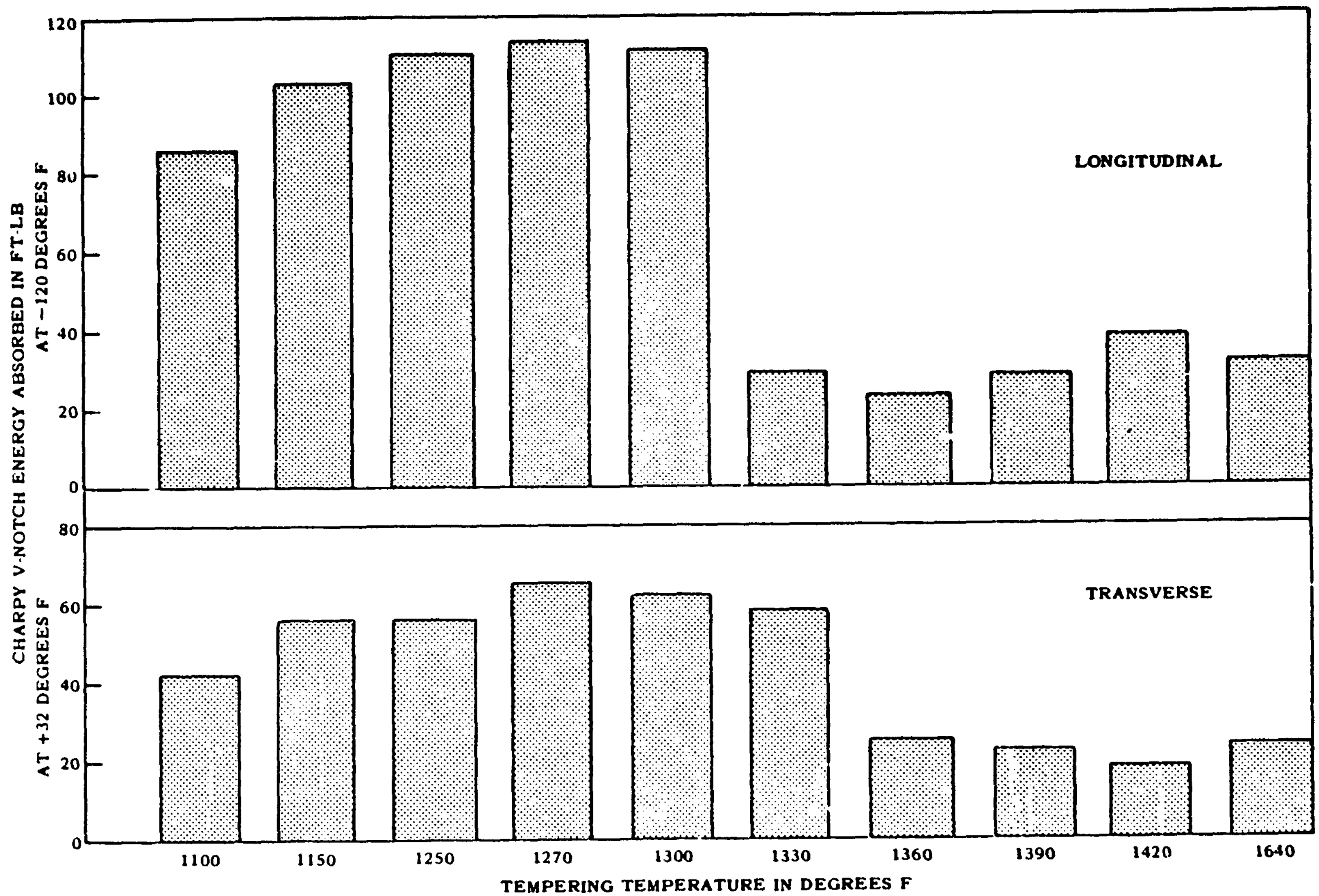


Figure 9 – Effects of Tempering Temperature on the Charpy V-Notch Energy Measured at +32 and −120 F

TABLE 1

Mechanical and Notch-Toughness Properties Obtained after Various Tempers

Tempering Temperature[1] deg F	Mechanical Properties							Charpy V-Notch Impact Properties										
	Compression	Tension				Property Ratios		Longitudinal			Transverse							
								−120 F			+32 F			NDT Temperature				
	0.2 Percent Yield Strength ksi	0.2 Percent Yield Strength ksi	Ultimate Tensile Strength ksi	Percent Elongation in 1 in.	Percent Reduction of Area	CYS/TYS	TYS/UTS	Energy ft-lb	Lateral Expansion mils	Percent Fiber	Energy ft-lb	Lateral Expansion mils	Percent Fiber	Temperature deg F	Energy ft-lb	Lateral Expansion mils	Percent Fiber
1100	-	113	124	19	60	-	0.911	86	48	0	42	33	100	−170	25	14	38
1150	114	104	116	22	63	1.096	0.896	103	66	0	56	34	100	−210	29	17	35
1250	100	91	105	27	78	1.099	0.867	110	62	0	56	44	100	−265	24	12	22
1270	92	85	103	28	77	1.082	0.825	114	74	0	65	55	100	−240	29	17	24
1300	86	82	104	27	76	1.049	0.788	119	76	0	62	59	100	−190	38	29	74
1330	88	81	105	27	76	1.086	0.771	29	19	12.2	58	53	100	−190	37	29	48
1360	107	86	165	19	58	1.244	0.521	23	41	13.7	25	15	100	−160	19	9	23
1390	123	96	172	18	60	1.281	0.558	28	13	14.0	22	12	100	−160	17	8	20
1420	162	127	195	15	60	1.276	0.651	38	16	32.9	18	10	100	−120	16	8	27
1640	161	148	200	14	45	1.088	0.740	31	12	31.6	23	8	100	−150	21	6	-

[1] Austenitized at 1640 F for 1/2 hour, water quenched; held at −120 F for 1 hour; tempered for 1 hour at indicated temperature, then water quenched.

TABLE 2
Comparison of Properties for a Given Strength Obtained by Tempering Above and Below the A_{c1} Temperature

Table 2a – For Same Ultimate Tensile Strengths

Property*	Temper		Temper		Temper	
	$>A_{c1}$(1360 F)	$<A_{c1}$ (900 F)	$>A_{c1}$(1390 F)	$<A_{c1}$ (800 F)	$>A_{c1}$(1420 F)	$<A_{c1}$ (400 F)
Ultimate Tensile Strength[1]	165	165	172	172	195	195
Tensile Yield Strength[1, 2]	86	152	96	157	127	162
Yield Strength/Tensile Strength	0.52	0.92	0.56	0.91	0.65	0.83
Percent Elongation in 1 in.	19	18	18	18	15	14
Percent Reduction in Area	58	66	60	64	60	61
Nil-Ductility Transition Temperature, F	-160	-110	-160	-115	-120	-165
CVT[3] at NDT[4] Temperature[3]	19	21	17	22	16	22
CVT[3] at +32 Degrees F[4]	25	25	22	25	18	25

Table 2b – For Same Tensile Yield Strengths

Property	Temper		Temper		Temper	
	$>A_{c1}$(1360 F)	$<A_{c1}$(1265 F)	$>A_{c1}$(1390 F)	$<A_{c1}$(1200 F)	$>A_{c1}$(1420 F)	$<A_{c1}$(1050 F)
Tensile Yield Strength[1, 2]	86	86	96	96	127	127
Ultimate Tensile Strength[1]	165	104	172	110	195	143
Yield Strength/Tensile Strength	0.52	0.83	0.56	0.87	0.65	0.89
Percent Elongation in 1 in.	19	28	18	25	15	19
Percent Reduction in Area	58	77	60	70	60	69
Nil-Ductility Transition Temperature, F	-160	-245	-160	-235	-120	-130
CVT[3] at NDT[4] Temperature	19	28	17	27	16	26
CVT[3] at +32 Degrees F	25	63	22	56	18	44

*Numerals indicate the following: 1 in KSI;
2 0.2 Percent Offset;
3 Transverse Charpy V-Notch Impact Energy Absorbed in ft-lb, and
4 Nil-Ductility Transition Temperature in Degrees F.

TABLE 3
Effects of Retempering at 1250 F after an Initial Temper above the Lower Critical Temperature

Treatment[1]		Mechanical Properties							Charpy V-Notch Impact Properties		
		Compression	Tension				Property Ratios		Transverse (+32 F)		
First Temper deg F	Second Temper deg F	0.2 Percent Yield Strength ksi	0.2 Percent Yield Strength ksi	Ultimate Tensile Strength ksi	Percent Elongation in 1 in.	Percent Reduction in Area	CYS/TYS	TYS/UTS	Energy ft-lb	Lateral Expansion mils	Percent Fiber
1360	-	107	86	165	19	58	1.24	0.521	25	15	100
1360	1250	76	74	96	30	79	1.03	0.771	54	51	100
1390	-	123	96	172	18	60	1.28	0.558	22	12	100
1390	1250	83	80	100	30	77	1.04	0.80	57	48	100
1420	-	162	127	195	15	60	1.276	0.651	18	10	100
1420	1250	98	92	107	26	77	1.065	0.860	70	58	100
1640	-	161	148	200	14	45	1.088	0.740	23	8	100
1640	1250	100	91	105	27	78	1.099	0.867	56	44	100

[1] Austenitized 1640 F for 1/2 hr, water quenched, -120 F for 1 hr; first temper for 1 hr, water quenched, second temper for 1 hr, water quenched.

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The effects of tempering above the critical transformation temperature on the microstructure, notch-toughness and mechanical properties of HY-80 steel was investigated. The formation of alpha ferrite and prime martensite after tempering above the lower critical temperature and quenching will result in mechanical properties that fall below the requirements of the HY-80 specification. Retempering to achieve minimum yield strengths may result in nonuniform yield strength distribution. The possibility of underbead cracking due to alloy segregation is also discussed.

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14. KEY WORDS	LINK A		LINK B		LINK C	
	ROLE	WT	ROLE	WT	ROLE	WT
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HY-80 Steel – Mechanical Properties						
HY-80 Steel – Metallurgical Properties						
HY-80 Steel – Microstructure						
HY-80 Steel – A_{c1} Temperature						
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