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NAVAL SHIP RESEARCH AND DEVELOPMENT CENTER

BETHESDA, MD 20034

EFFECTS OF MICROSTRUCTURE, COMPOSITION, AND STRENGTH ON THE STRENGTH-DIFFERENTIAL PHENOMENON OBSERVED IN HY-80 STEEL

by

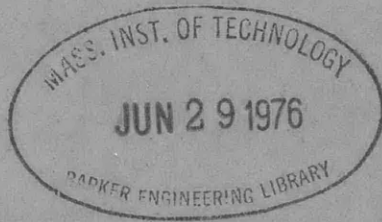
M. L. Salive and A. R. Willner

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STRUCTURES DEPARTMENT
RESEARCH AND DEVELOPMENT REPORT



October 1971

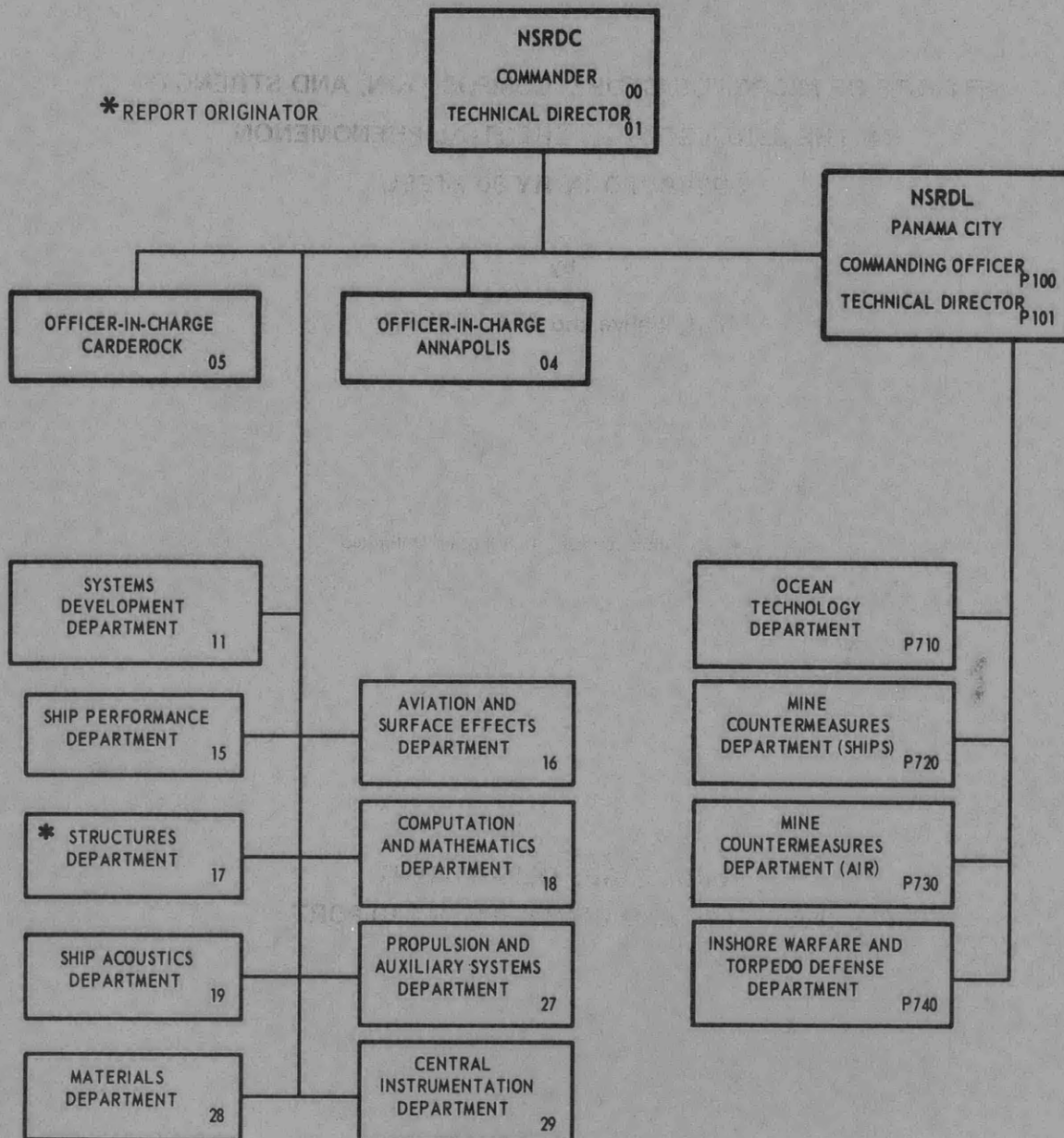
Report 3701

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Naval Ship Research and Development Center
Bethesda, Md. 20034

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DEPARTMENT OF THE NAVY
NAVAL SHIP RESEARCH AND DEVELOPMENT CENTER
BETHESDA, MD 20034

EFFECTS OF MICROSTRUCTURE, COMPOSITION, AND STRENGTH
ON THE STRENGTH-DIFFERENTIAL PHENOMENON
OBSERVED IN HY-80 STEEL



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ABSTRACT

Steel from 22 heats of low-carbon Ni-Cr-Mo steel (MIL-S-16216G and ASTM A543-65) were heat treated to study the effects on the strength-differential effect and the difference between tensile and compressive yield strength of (1) commercial variation in composition and inclusion content, (2) variation in microstructure such as prior austenitic grain size and the relative amount of isothermally produced ferrite or bainite in a tempered martensitic matrix, and (3) the observed variation in strength obtained after a 1-hour 1150 F temper followed by a water quench to prevent embrittlement while cooling from the tempering temperature.

The difference between the tensile and compressive yield strength, sometimes called the strength differential (S-D) effect, was observed in this study to be at least 5 percent of the tensile yield strength. Data are cited to show that in the low-carbon Ni-Cr-Mo steels studied here, the S-D effect observed was a relatively constant percent of the tensile yield strength, and was markedly structure-sensitive to prior austenitic grain size, microstructural constituents, tempering temperature, type and distribution of carbides formed during tempering, and tempering slightly above the lower critical (A_{C1}) temperature.

ADMINISTRATIVE INFORMATION

This work was performed under the sponsorship of the Naval Ship Systems Command (NAVSHIPS) as part of Research and Development Project SF 35.422.212, Program Element 62512N.

INTRODUCTION

Hirth and Cohen¹ have called attention to the strength-differential phenomenon* in martensitic steels wherein the strength of such steels is notably higher in compression than in tension. Recently Leslie and Sober observed that martensitic 4300 series steels are significantly stronger in uniaxial compression than in uniaxial tension.² Lessels has observed a similar effect for 1035 and 1045 steels.³ Hirth and Cohen have presented a partial explanation for the strength-differential (S-D) effect in as-quenched martensitic steel based on the interstitial carbon atoms straining the body centered iron from a cubic to a tetragonal lattice. They attribute the S-D effect to the difference between the nonlinear stress-strain curves in tension and compression and demonstrate that their proposed nonlinear model can account for a 3 to 6 percent strength-differential.

The S-D effect is of practical importance to the materials engineer as well as to the metallurgist, since this effect has been observed for brittle materials such as glass, ceramic, and cast iron, as well as for ductile metals.²⁻⁵ The design engineer becomes vitally concerned with the S-D effect when he is called upon to design weight critical structures subjected to compressive stresses. Perhaps the most obvious example of this is in the design of oceanographic devices where the safety factor must be held to a minimum so as to reduce the amount of dead weight that has to be counterbalanced by additional buoyancy material or floatation spheres. The tremendous external hydrostatic pressures imposed on deep diving oceanographic vehicles require the use of materials with precisely known compressive yield strengths loaded to as high a stress as possible in order to

¹References are listed on page 48.

*The difference between tensile and compressive uniaxial yield strength. The percent S-D effect is $((CYS-TYS)/TYS) \times 100$, and $\Delta\sigma_{S-D}$ is $(CYS-TYS)$.

minimize the dead weight of the pressure resistant hull and to maximize the effective payload in men and instrumentation.

The design engineers at the Naval Ship Research and Development Center (NSRDC) have long been aware of the S-D effect and as a result regularly measure the uniaxial compressive properties in the direction of the principal stresses in their structural models of oceanographic vehicles and submarines. Compression testing of as-received materials is essential, since all structural materials are regularly purchased on the basis of their tensile properties. The S-D effect can introduce enough difference between the calculated and measured hydrostatic collapse pressure of a model to obscure the relative effect of small changes in structural design elements such as frame spacing, sphericity, or the interaction of penetration reinforcements around viewing ports and hatches.

As a result of Navy interest in the S-D effect, certain studies made on the effects of metallurgical variables such as microstructure and strength level included comparative data on the uniaxial tensile and uniaxial compressive yield strength of HY-80 steel.⁴⁻⁵ The pertinent results of these studies will be compared to the predicted results of the nonlinear model of the solute/dislocation interaction hypothesized by Hirth and Cohen to see if the assumptions upon which the hypothesis is based are valid.

BACKGROUND

The work of Hirth and Cohen was stimulated by Leslie and Sober's work on the effects of 0.12, 0.21, 0.34 and 0.41 percent carbon additions on the strength of as-hardened martensitic AISI 4300 series steels in the brine quenched condition.^{1,2} Figure 1 gives a plot of the S-D effect ($\Delta\sigma_{S-D}$) versus carbon content and a plot of the ratio of the magnitude of the S-D effect to the tensile yield strength (σ_{TYS}) for various carbon contents.¹ The solid line represents the experimental values and the hatched area gives the region in which the true strength differentials would lie, based on the 1/3 to 1/2 correction to $\Delta\sigma_{S-D}$ that Rauch and Leslie reportedly recommend be applied to this data.¹

It should be pointed out that Leslie and Sober heat-treated finished specimens in evacuated quartz capsules which were crushed upon contact with the brine quench.² The specimens were reported not to have been distorted, although the higher carbon specimens increased in length slightly. They also reported that about 1/3 of the 4340 specimens had to be discarded since they were cracked when they were taken from the liquid nitrogen bath. This cracking is frequently observed for the higher carbon steel and is attributed to the lowering of the M_S temperature as the C content is raised. The lower M_S eliminates the possibility of self tempering of the fresh martensite to relieve residual stresses in the brittle, non-ductile as-quenched martensite. It is generally believed that the lower carbon martensites are not only inherently more ductile, but that they self temper somewhat due to the higher temperature at which they form.

Leslie and Sober's compression tests were reported to have been made using a subpress without either a self-aligning swivel head or lubrication between the ends of the specimen and the anvils, as recommended by the ASTM. In addition, they used a deflectometer to measure strains. As is usually the case when using deflectometers, instead of an averaging compressometer with a fixed gauge length, the apparent modulus of elasticity of the steel specimen was $\leq 20 \times 10^6$ psi instead of the expected 30×10^6 psi; this frequently leads to mistakes when taking the 0.2 percent offset yield strength, and may account for the need of a correction factor. Tension tests were made using a 1 inch gage length extensometer. The uncorrected data show that as the carbon content of the martensite is lowered, the S-D effect decreases. Other data is cited to show that the S-D effect (a) is typically 10 percent;⁶ (b) is reported to go to zero in as-quenched 0.025 percent C ferrite;⁷ (c) persists beyond

yielding out to about 3 percent strain; (d) is characteristically smaller for a bainitic structure than for martensitic structures; and (e) does not vanish until a relatively high tempering temperature (> 1250 F) is reached.²

In addition to the data on as-quenched 4300 steels, Leslie and Sober gave the following data on the effects of tempering for 1 hour on the S-D effect in 4330 steel:

Tempering Temperature	σ_{CYS}	Strength, ksi σ_{TYS}	$\Delta\sigma_{S-D}$
72 F	271	220	51
400 F	237	210	27
800 F	200	180	20
1200 F	125	115	10

It is uncertain whether the 1/3 to 1/2 correction factor should be applied to these values, but it is most likely necessary since these data are from the same set of experiments. The hatched areas on Figure 1c represent the effect of applying this correction.

Hirth and Cohen offer and discuss six tentative explanations of the observed S-D effect, most of which they reject. Briefly, their explanations and observations are as follows:

1. **MICROCRACKS** – rejected because while they may be more detrimental to σ_{TYS} than to σ_{CYS} in some cases, they are not detrimental for fine-grained fracture tough steels.

2. **RESIDUAL STRESSES** – caused by the volume expansion that accompanies the austenite to martensite transformation and by the thermal gradients resulting during quenching; rejected as a source of the S-D effect since:

(a) they must be balanced by equal and opposite forces to maintain micro and macro equilibrium;

(b) the S-D effect persists out to about 3 percent strain where it should have been wiped out by that amount of plastic strain; and

(c) it persists even after 25 percent reduction in thickness by rolling.

3. **RETAINED AUSTENITE** – transforms more readily to martensite under tensile loading and the accompanying volume expansion could lead to premature yielding; this theory must be rejected since:

(a) the Leslie-Sober specimens were held at liquid nitrogen temperature and there was essentially no retained austenite as verified by X-ray diffraction;

(b) the S-D effect is observed even after tempering at temperatures high enough to have transformed any remaining austenite; and

(c) it persists after 25 percent deformation, which is sufficient to transform any retained austenite to below detectable levels.

4. **INTERNAL BAUSCHINGER EFFECT** – this hypothesis, which is based on the residual stresses set up by the volume expansion accompanying the austenite to martensite transformation and the resultant plastic compression of the martensite plates or needles by the surrounding material, is rejected by Hirth and Cohen for two reasons:

(a) it requires selective orientation of the martensite relative to the axis of testing, and this is unlikely even within a single parent grain, since there are 24 different crystallographic orientations that the martensite can form along in the parent grain, which itself is usually randomly oriented.

(b) if there were some S-D effect in one of the principal directions of a hardened plate due to a Bauschinger effect, it should be reversed or appreciably lower in the transverse direction, and such is not the case.

5. VOLUME EXPANSION – is rejected as plausible but difficult to evaluate since it would be dependent upon whether the actual expansion accompanying plastic strain is sufficient to account for the S-D effect and whether it takes place during a rate controlling event. As an example, Hirth and Cohen indicate that the sign (\pm) of normal stress across the glide plane may affect the Peierls stress.

6. SOLUTE/DISLOCATION INTERACTION IN SOLID SOLUTION – this is the hypothesis selected by Hirth and Cohen for their model and based on the distortion caused to the iron lattice when the austenite to martensite transformation traps most of the available carbon atoms in solution. This has two possible effects:

(a) the Peierls stress for movement of dislocations across the glide plane is dependent on the sign of the normal stress, with the Peierls stress increasing as σ_n decreases and eventually becomes negative, and

(b) the tetragonal distortion of the iron lattice around the trapped carbon atom causes nonlinear elastic strains which result in the plastic force-displacement relations not being the same in tension and compression, thus causing the S-D effect. Hirth and Cohen state that the Peierls stress is not controlling since the S-D effect is not as temperature dependent as would be expected if it did control; observed temperature dependence is related instead to obstacle spacing and dislocation density. The Hirth-Cohen model is based on the distortions caused by the interstitial carbon atom and determines the order of magnitude of the contribution of the nonlinear lattice distortions to the S-D effect.

Hirth and Cohen derive equations based on this last theory to calculate that the nonlinear interaction energy difference associated with the S-D effect is of the order of 4 percent of the carbon/dislocation interaction energy for various values of strain associated with a dislocation strain field and for an applied strain $\epsilon_a = \pm 0.01$. They state that they expect the S-D effect to persist at a decreased level in tempered martensite and tempered bainite, since the interaction is then associated with carbon atmospheres adsorbed at dislocations. In addition, they expect linear dependence of flow stresses on both solute (carbon) concentration and the solute/dislocation energy for drag of Cottrell, Snoek or core atmospheres and for dislocation motion without drag effects in concentrated solid solutions or solutions exhibiting order. They state that the solid solution strengthening is expected to be proportional to the carbon/dislocation interaction energy as well as some power of the carbon concentration.

Additional data are available on the S-D effect in an aluminum killed lower carbon, 0.16 percent, quenched and tempered HY-80 steel of the following composition: 0.16 C, 0.30 Mn, 0.009 P, 0.017 S, 0.25 Si, 2.78 Ni, 1.52 Cr, 0.36 Mo, 0.058 Cu, balance Fe.^{4,5} Figure 2 is taken from References 4 and 5 and shows the effects of tempering temperature on the S-D effect in quenched and tempered martensite. This material was austenitized at 1640 F for 1/2 hour. All tempering was performed on as-quenched martensite that had been held for 1 hour at –110F in dry ice and acetone. Tempering was for 1 hour at the indicated temperature and was followed by

a water quench from the tempering temperature. The lightly dashed line on Figure 8a is for the same steel after austenitizing at 2000 F for 1 hour.

Figure 3 shows similar data for this material⁴ after austenitizing at 2000 F for 1 hour, transferring to 1640 F for 5 minutes and then either directly brine quenching or isothermally treating to produce some bainite or ferrite and then water quenching. The material was given a low temperature treatment at -110 F for 1 hour immediately after quenching. Tempering was for 1 hour at the indicated temperature and was followed by a brine quench from the tempering temperature. All the data in Figures 2 and 3 were obtained using self-aligning test fixtures and averaging, 1 inch gage length compressometers and extensometers.

EXPERIMENTAL PROCEDURES

SUMMARY

The general test program followed in this study can be summarized as follows. Test coupons were collected from commercially produced plates made by the two major producers of a low carbon, nickel-chrome molybdenum steel.* These coupons were selected to represent the broad range of chemical composition to which this steel is melted in commercial practice. These coupons were slabbed to convenient size, analyzed for chemical composition, heat treated, and tested. All heat treating and testing was performed on 5/8 x 2 x 5 inch drop weight blanks heat treated so as to eliminate the effects of variation in hardenability, plate thickness, and mill heat treatment. Isothermal heat treatments were used to produce controlled microstructures, and a uniform tempering temperature of 1150 F, slightly above the minimum value permitted by the specification, was used. The isothermal holding times and temperature were tentatively selected to produce approximately 25 ± 10 percent of each non-martensitic product, bainite or ferrite, for an average composition. Microstructures produced by a given heat treat were measured for each coupon using microspecimens taken from the broken drop weight test specimens; mechanical properties were measured on these same test specimens.

MATERIALS

Coupons were collected over a period of years from plates made during actual mill runs at the plants of the two major producers of this steel and represent only a small fraction of heats melted during this period. The coupons were obtained after final heat treatment from the end of the plate when the plate was cropped to ordered size. All coupons were cut from material adjacent to the ordered plate. The actual coupons selected were chosen by the investigator on the basis of the producer's chemical analysis to represent a fairly broad range of the chemical compositions used for this steel in commercial mill practice. A total of 27 coupons representing 22 heats were collected; 16 coupons were from one producer and 11 coupons from the other. Table 1 summarizes the heat numbers, mechanical properties and chemical compositions reported by the producers; in addition, the table lists the identification numbers assigned to each coupon. Coupons made by the largest producer have a number starting with an X and coupons made by the other producer have been assigned a number prefixed with a Y. To obtain the maximum use of each coupon and to make specimen

*Commonly called HY-80 steel and made in accordance with Military Specification MIL-S-16216 G and ASTM Specification A543-65.

location as uniform as possible for all specimens, specimens were taken from adjacent to the original plate surfaces (see Figure 4). In fact, to minimize machining costs, the original plate surface was made one of the 2 × 5 inch surfaces of the drop weight specimen blank. The specimen blanks were 5/8 × 2 × 5 inch with the 2 inch dimension in the direction of plate roll and the 5 inch dimension parallel to the plate surface and perpendicular to the direction of plate roll (see Figure 5). This is the size of specimen required for subsize drop weight Nil Ductility Transition (NDT) temperature tests in ASTM Method E208. Each specimen blank was coded on the end with the appropriate number, location, test direction, and specimen number. These specimen blocks were used for drop weight tests; after that the broken specimens were used to make mechanical property test specimens.

HEAT TREATING PROCEDURES

The 5/8 × 2 × 5 inch drop weight specimen blanks were prepared for heat treatment by (1) welding handling loops onto one of the 5/8 × 2 inch faces, (2) depositing a brittle crack starter bead on the 2 × 5 inch surface of the specimen that represented the original plate surface of the specimen using a 3/16 inch diameter Hardex *N* hardfacing electrode, (3) normalizing the specimens by heating to 1650 F, holding for 1 hour and then air-cooling, and (4) sandblasting the specimen prior to final heat treatment.

The final heat treatment of the drop weight specimen blanks consisted of austenitizing in a neutral salt bath at 1640 F for 1/2 hour and then either water quenching in a brine solution or quenching into an isothermal neutral salt bath, holding for a prescribed time and then brine quenching. After quenching, all specimens were immediately given a low temperature (about –110 F) treatment in a bath made of acetone and chunks of dry ice; all specimen blanks were given at least 1/2 hour in this bath to ensure that they came down to temperature and that all retained austenite was transformed to martensite by the treatment. All specimens were subsequently tempered in a circulating air furnace at 1150 F for 1 hour at temperature and then water quenched in a brine solution (*BQ*). In a few cases, where sufficient material was available, a few sets of drop weight specimen blanks were austenitized at 2000 F for 1 hour in a circulating air furnace, then transferred to the 1640 F salt bath for 5 minutes, minimum, to standardize the temperature differential between the isothermal bath and the specimen prior to immersion. All furnace and bath temperatures are manually controlled using potentiometers to within ±5 F. Specimens were heat treated by suspending them by their loop from a fixture that allowed them to be treated as a group. This fixture held the specimens 1 1/2 inches apart and suspended them vertically in the salt baths to permit circulation of the salt.

Fully Quenched (Martensitic)

Specimen blanks that were fully quenched to form a fine prior austenitic grain size to ultimately give a tempered martensitic microstructure were treated as follows: 1640 F – 1/2 hour – *BQ*; –110 F – 1/2 hour; temper 1150 F – 1 hour – *BQ*. Specimen blanks from coupons X-12, -13, -15 through -18 which were quenched from a coarse prior austenitic size were treated as follows: 2000 F – 1 hour – 1640 F 5 minute (minimum) – *BQ*; –110 F – 1/2 hour; temper 1150 F – 1 hour – *BQ*.

875 F Isothermal Treatment (Semi-Bainitic)

Specimen blanks that were treated isothermally at 875 F to ultimately give a microstructure consisting of a mixture of tempered bainite and tempered martensite were treated as follows: 1640 F – 1/2 hour – transferred to 875 F – held 152 (or 1600) seconds – *BQ*; –110 F – 1/2 hour; temper 1150 F – 1 hour – *BQ*. Additional specimen blanks available from coupons X-12, -13, -15, -16, -17, and -18 were treated as follows: 2000 F – 1 hour – transferred to 1640 F held 5 minutes (minimum) – transferred to 875 F – held 1600 seconds – *BQ*; –110 F – 1/2 hour; 1150 F – 1 hour – *BQ*.

1200 F Isothermal Treatment (Semi-Ferritic)

Specimen blanks from all coupons were isothermally treated at 1200 F to ultimately give a microstructure of tempered martensite and tempered ferrite. In several cases a little pearlite was found to have been formed during the isothermal treatment. These coupons were treated as follows: 1640 F – 1/2 hour – transferred to 1200 F – held for 3350 seconds (extra sets of specimen blocks from coupons X-1, -2, -12, -13, -15 through -18 and Y-2, -4, and -11 were also held for 8500 seconds) – *BQ*; –110 F – 1/2 hour; 1150 F – 1 hour – *BQ*. There was insufficient material to give any specimens with a coarse austenitic grain size the 1200 F isothermal treatment.

TESTING PROCEDURES

In general an attempt was made to adhere to standard ASTM testing procedures. In a few cases it was necessary to make slight deviations from standard practice to take into account the specific conditions of these tests; any such deviations will be indicated below.

Drop Weight NDT Tests

All drop weight tests were performed in accordance with ASTM Method E 208-66T, using the type P-3 specimen which is 5/8 x 2 x 5 inch. The specimens were supported as a simple beam and subjected to an impact load of 259 ft-lb produced by dropping a 223 lb weight 1 foot 2 inches; this drop energy was found to consistently crack the crack starter weld bead and to cause the specimen to consistently contact the anvil stop.

As stated previously, the specimens were prepared with one, as rolled, 2 x 5 inch face. The Hardex-*N* crack starter bead was deposited in the center of this face prior to heat treatment of the specimen so that the heat-affected zone created by deposition of the weld bead would not interfere with the correct measurement of the NDT temperature of the base material. After final heat treatment, the crack starter weld beads were notched as required in ASTM E 208, using a thin abrasive disk. The combination of the isothermal treatments and the subsequent 1150 F temper was not found to prevent proper functioning of the crack starter bead; upon impact, a good crack formed in the bead in every case.

A group of four or five, usually five, specimens were given a selected heat treatment and used to determine the NDT temperature. Since tests were conducted on a known material with a known thermal history, the NDT temperature was determined within 10 F by three to five tests; the usual number being four. Prior to testing, each specimen was placed in a large insulated liquid, isopentane, low temperature bath with at least 1 inch of

liquid all around the specimen. A thermocouple embedded in a dummy specimen was attached to an automatic recorder and used to control the bath temperature. The bath temperature was maintained for a period of at least 45 minutes prior to test by manually stirring the bath and adding dry ice or liquid nitrogen as required. Gloves were used to handle the specimens by the heat treating loop welded to the end (5/8 x 2 inch face). Interpretation of the specimen fracture as to "Break" or "No Break" was as recommended in ASTM E 208. The only "No-Test" specimens encountered in these tests were with a few specimens that were cut a little thin, less than the recommended 0.62 ± 0.02 inch thickness. These specimens were not included in the measurement of the NDT temperature.

Tensile Tests

All tensile tests were performed in accordance with ASTM Standard E 8-65T, Tension Testing of Metallic Materials. A small size specimen, 0.250 inch diameter with a 1 inch gage length, proportioned to the standard Type 1, 0.505 inch diameter tension test specimen was used. The specimens were taken parallel to the original direction of plate roll; thus the overall length of the specimen was limited to 2 inch since this is the size of the P-3 drop weight specimen in that direction (see Figure 6). The 2 inch overall length was achieved by slightly reducing the length of the threaded ends of the specimen.

The specimens were tested in hydraulically loaded universal testing machines calibrated in accordance with ASTM Standard E 4-64, Verification of Testing Machines, using proving rings. The testing machines were found to indicate the correct load within ± 0.5 percent in the loading ranges used in these tests. Self aligning, spherical bearing devices were used to ensure axial loading of the test specimens. Load-strain curves were recorded for all specimens using an averaging, 1 inch gage length, microformer strain indicator (extensometer) and an autographic recording device calibrated over the range of the expected yield. Specimens were tested at a speed of 0.002 inch/inch/minute or less, up to and slightly beyond the yield point. The 0.2 percent offset method was used to determine the yield strength. The values reported are the average of two specimens.

Compression Tests

All compression tests were performed in accordance with ASTM Standard E 9-61. The specimens tested were 1/2 inch diameter by 2 inch long cylinders made with special care to ensure that the ends were parallel to each other within 0.002 inch and perpendicular to the long axis of the specimen within 0.005 inch. The specimens were taken parallel to the original direction of plate; thus they were machined parallel to the 2 inch dimension of the drop weight specimen they were cut from (see Figure 6).

The specimens were tested using hydraulically loaded testing machines calibrated in accordance with ASTM Standard E 4-64, Verification of Testing Machines. The testing machines were verified using proving rings and found to indicate the correct load within ± 0.5 percent in the loading ranges used in these tests.

All compression tests were run using a subpress with a spherical seated bearing block at the end of its shaft to ensure full contact with the end of the specimen and to ensure true axial loading. Hardened steel bearing blocks with smooth, ground, parallel faces were used in contact with the ends of specimen to prevent possible damage to the faces of the subpress. The ends of the specimen were greased with lubriplate to eliminate end restraints. Because of the shape of the specimen it was not necessary to use a jig to prevent lateral buckling.

Load-strain curves were recorded for all specimens using an averaging, 1 inch gage length, microformer strain indicator (compressometer) and an autographic recording device calibrated over the range of the expected yield. Specimens were tested at a speed of 0.002 inch per inch per minute or less up to and slightly beyond the yield point. The 0.2 percent offset method was used to determine the yield strength. The values reported are the average of two specimens.

METALLOGRAPHIC PROCEDURES

Metallographic specimens were cut from the undeformed areas of the drop weight test specimen blank and far enough from the crack starter bead to be sure that no alloying elements had diffused into the specimen. Longitudinal and transverse specimens were prepared and mounted in transoptic mounting resin. After the specimens were mounted, 1/8 inch of specimen surface was machined off in a lathe under coolant using a sharp cutting tool. The surface was progressively ground using 180-, 240-, and 600-grit silicon carbide papers. Final polishing was performed in two stages using 60-cycle automatic vibratory polishers. The semifinal polishing bowl was covered with bleached silk cloth and a slurry of Linde A, distilled water and aerosol wetting agent. The final polishing bowl was covered with Gamal cloth and a slurry of Linde B, distilled water, and aerosol. All specimens were lightly etched in 2 percent Nital and repolished several times during the final stages of polishing to ensure the removal of any smeared structure and correct development of the microstructure.

The inclusion content of the specimens was measured in both the longitudinal and transverse, through-the-thickness planes in accordance with ASTM Standard E 45-63, Determining the Inclusion Content of Steel. Microscopic Method B was used to determine the length of the longest inclusion, the number and average length of all inclusions over 0.005 inch long and the background rating including all inclusions less than 0.005 inch long. The vibratory polishing technique was found to leave the inclusions intact and in the specimen.

The prior austenitic grain size was measured on fully-quenched specimens taken from the quenched and tempered fully martensitic specimens in accordance with ASTM Standard E 112-63, Estimating the Average Grain Size of Metal, using the intercept or Heyn procedure. The prior austenitic grain size was revealed by the use of the etching reagents described in Appendix A. Comparisons made on several different specimens from different melts showed that both etches indicated the same prior austenitic grain size.

The percent of isothermal transformation product resulting from the isothermal treatments was measured for each treatment. The microspecimens were each etched in three reagents. Between each of the reagents the specimens were rinsed and dried. The triple etch consisted of (1) a saturated picric acid solution to develop the prior austenitic grain boundaries, (2) a 1 percent nital solution to develop the structure in the areas of tempered martensite, and (3) a 20 percent solution of anhydrous sodium metabisulfite for staining (darkening) the tempered martensite areas. The last etch leaves the areas of the tempered bainite or ferrite white, in sharp contrast to the darker areas of stained tempered martensite. The details of these etching solutions and how they were used are given in Appendix B. The percent of transformation product was estimated by superimposing a grid on the specimen and determining the percentage of the grid intersections that fell over each microconstituent. Because of the finely banded nature of the transformation product and the high magnification (500X/1000X) needed to resolve the structure, extreme care was taken to systematically traverse the specimen and read at least 10 areas so as to get a representative average value to report.

EXPERIMENTAL RESULTS

CHEMICAL COMPOSITION

The actual chemical compositions of the coupons tested in this study are listed in Table 2. These analyses were made by a commercial testing laboratory on material located adjacent to the original plate surface of the coupon. The material analyzed came from the top 5/8 inch of the plate but did not include any heat treating scale or decarburized plate surface which would have influenced the analysis.

Wet chemical analyses were performed for percent Carbon (C), manganese (Mn), phosphorus (P), silicon (Si), nickel (Ni), chrome (Cr), molybdenum (Mo), copper (Cu), and acid soluble and total aluminum (Al).

Spectrographic analysis was performed to determine the percent present of the following trace elements: vanadium (V), lead (Pb), tin (Sn), magnesium (Mg), cobalt (Co) and titanium (Ti). No values are reported for Pb, Sn and Mg in Table 2 since they were not found to be present in detectable amounts. After several samples had been analyzed it was found that V and Co were present in negligible amounts and therefore, analysis for these elements was discontinued. The producers informed the author that the Co was a tramp element introduced by the nickel addition. The Co was left in with the nickel since it was too difficult to remove and since they found that it had essentially no effect on the hardenability or properties of the steel. The V was also present only as a tramp element but it came from a variety of sources that went into the makeup of a given heat of steel.

Gaseous analysis was performed for oxygen (O₂), hydrogen (H₂), and nitrogen (N₂). O₂ and N₂ are reported as a percent and H₂ as parts per million (ppm) in Table 2.

A review of Table 2 shows that in general the range of the alloying elements found in the samples selected for this study almost completely cover the full composition range permitted for nickel-chrome-molybdenum steels such as ASTM A543 and HY-80. The only element that falls outside the combined range for the two steels is Mn; coupons Y-8 and Y-10 have 0.49 percent Mn. This is only 0.04 percent over the 0.45 percent (0.40 upper limit plus the allowance of 0.05 percent variation over the upper limit) maximum allowed for these steels.

METALLOGRAPHIC ANALYSIS

The results of the metallographic analysis of coupons used in this study are given in Tables 3, 4, and 5. Table 3 presents the results of the inclusion analysis of these coupons at the mid-thickness and adjacent to the surface of each coupon in both the longitudinal and transverse directions on planes perpendicular to the plate surface through the thickness of the plate. The reporting method recommended in ASTM E 45-63 is used to describe the inclusion content. The length of the longest inclusion at 100X is reported in units equivalent to 0.005 inch on the specimen along with a superscript describing whether it is grouped (g), very disconnected (vd) or disconnected (d). The average length of all inclusions over one unit long and excluding the longest is reported with a superscript denoting the number of inclusions averaged. The background rating A, B, C or D corresponds to the sample photographs in E45 with A being a rather clean steel and D a rather dirty steel. Figures 7 and 8 are photomicrographs (100X) of representative samples and will give the reader a better understanding of the significance of the data reported in Table 3.

The prior austenitic micro-grain size of each coupon is reported in Table 4. The grain size is reported for the as-received coupon and for the coupons after heat treatment. This table shows that the coupons austenitized at 1640F had a slightly finer prior austenitic grain size than as-received coupons, and a slightly coarser grain size after the 2000F treatment.

The percent of isothermal transformation products produced by the various isothermal treatments are summarized for each coupon in Table 5. The large variation in percent transformation product resulting from a given heat treatment reflects the large range in chemical composition and hardenability of these coupons. The high hardenability, rich composition coupons had the least transformation take place for a given isothermal treatment. Photomicrographs of typical areas of each coupon for each isothermal treatment are given in Appendix C. The dark areas are tempered martensite that has been darkened as described in Appendix B and the light areas are tempered isothermal products.

MECHANICAL PROPERTY TEST RESULTS

Table 6 summarizes the mechanical property test results for each heat treatment; the corresponding microstructures are also included for the reader's convenience. The table lists the longitudinal tensile and compressive properties, and the NDT temperature. All strength values reported are the average of two specimens.

DISCUSSION

A comparison of Tables 1 and 2 shows that there are some significant differences between the producer's reported chemical composition of the heats and the actual composition of the coupons. The biggest differences are in C, Mn, P and S with the coupon values generally higher than the producer's values. These variations are slightly more than what might be expected from normal variation in analytical results and from variation caused by the differences in the position of the sample in the heat of steel, and point up the need to check the actual composition of coupons being evaluated in laboratory experiments rather than to rely on the producer's data. Table 2 shows a good compositional range for the usual alloying elements and a fair spread for most of the rest of them.

Table 3 shows the mid-thickness and surface inclusion content of the coupons. In general, the material adjacent to the surface is cleaner than at the mid-thickness of the coupon. Even though a good range of inclusion sizes is represented in these samples, it must be pointed out that no large continuous inclusions were found.

Table 4 compares the prior austenitic grain size of the as-received material at mid-thickness to the prior austenitic, gamma grain size that resulted from the austenitizing treatments used in this study. While the as-received gamma grain size was fairly coarse, ranging from ASTM 8 to 4, the treatments used in this study bracketed an equally wide range in gamma grain size of ASTM 9.5 to 3.0.

The isothermal holding times referred to in Tables 5 and 6 are time in the bath, not actual holding times at bath temperature. Appendix D gives a comparison of the actual time-temperature history of 5/8 x 2 x 5 inch drop weight specimens when they are transferred from 1640F into either 875 or 1200F salt baths and compares it to the cooling curves predicted using the method of Sinnott and Shyne.⁸ The mid-thickness of the drop weight specimen is within 20F of 875F in about 120 seconds, a considerable portion of the 152-second treatment at 875F but a reasonable fraction of the 1600-second treatment. The data indicate that the mid-thickness of the specimens is within 20F of 1200F in about 90 seconds, which is negligible compared to the 3350- and 8500-second holding times used at 1200F. The important matter concerning holding times is that they produced a reasonable range in the amount of microconstituents.

Table 5 reports the amount of bainite produced during the 875F isothermal treatments and the amount of ferrite and pearlite produced by the 1200F treatments. The balance of the austenite that was untransformed

at the end of the isothermal treatment was completely transformed to martensite by the brine-water quench and -110F treatment. The amount of bainite produced by the 875F treatment ranged from 0 to 76 percent and varied with the hardenability of the particular steel and the isothermal holding time. As would be expected, the steels with the highest hardenabilities transformed the least. A similar effect was observed for the 1200F isothermal treatment, the amount of ferrite ranging from 0 to 69 percent. Only four of the 1200F treatments produced measurable amounts of pearlite, ranging from 4 to 32 percent; these steels were low hardenability, low Ni, and low Mn compositions. The fact that only four samples had pearlite in the microstructure must be considered during the subsequent discussion of the effects of microstructure on the S-D effect. There are so few points that not much confidence should be placed on the estimated effect of pearlite as opposed to the other microconstituents. It must be pointed out that one normally would not find pearlite in the microstructure of a low-carbon, Ni-Cr-Mo alloy steel such as HY-80 unless something were grossly wrong with the heat treatment and composition. Any discussion of the effect of pearlite is of academic rather than practical interest.

Table 6 gives a summary of the microstructure, tempered-bainite, -ferrite, -pearlite, and -martensite, longitudinal mechanical properties, and the drop weight NDT temperature resulting from the various heat treatments. Because of the uniform 1150F temper, the strengths reported here are more characteristic of an HY-100 steel rather than an HY-80 steel. The variations in strength reflect the variety of chemistries studied and the different initial hardnesses, the slightly different responses to tempering, and the different microstructures. The mechanical property data in Table 6, the chemistry data in Table 2, the inclusion content data in Table 3, the prior austenitic grain size data in Table 4, and the microconstituent data in Table 5 were combined to form the set which is the basis of this analysis. The result is that the 115 data points represent the 27 different coupons which constitute the sample used in this analysis.

EFFECT OF MICROSTRUCTURE ON STRENGTH DIFFERENTIAL PHENOMENON

The difference in uniaxial tensile and compressive yield strength in martensitic steels has been attributed to the difference in the nonlinear effects in tension and compression caused by the carbon atoms trapped in the martensite. Hirth and Cohen's model of the nonlinear stress-strain relation is based on a number of assumptions.¹ They assume the following:

1. The nonlinear effects are due to carbon atoms trapped in the interstices of the iron lattice and are not due to the substitutional alloying effects
2. The solid solution strengthening caused by the carbon is proportional to the carbon/dislocation interaction energy as well as some power of the carbon concentration. (It should be noted that Hirth and Cohen make no attempt to proportion this effect to carbon content in either their model or their sample calculation with 4340 steel. In fact, their only attempt to explain carbon concentration effects was to point out that the uncorrected $\Delta\sigma_{S-D}$ data of Leslie and Sober indicated that the S-D effect appears to increase linearly with carbon content.)
3. It is safe to neglect lattice forces and strains of the Fe/C bond between the interstitial C atom and its two nearest neighbors.
4. The lattice strains existing in the eight nearest Fe/Fe pairs will represent the nonlinear shift in the stress-strain relationship to within at least the correct order of magnitude.

5. The C atom causes a large dipole distortion which results in an 8 percent compressive strain in the bonds between the eight surrounding Fe/Fe atom pairs.

6. The nonlinear normal forces along the (111) direction are adequately estimated from cited stress-strain curves for (111) iron whiskers.

7. It is safe to disregard the straining effects of substitutional alloy additions on the basic iron lattice structure, since they would only add to the effect.

8. The S-D effect is maximized in the as-quenched martensitic structure and will disappear or persist at a decreased level in tempered martensite and bainite, the persistence being attributed to C atmospheres adsorbed at dislocations.

Assumptions 5 and 6 are based on experimental data and need not be questioned; assumptions 3, 4, and 7 are necessary to the development of the model with a minimum of complicating factors. Others of these assumptions 1, 2, and 8 can be examined in light of available experimental data, and it is these latter assumptions to which we turn to evaluate Hirth and Cohen's model of the S-D effect.

Leslie and Sober's data for as-quenched 4300 series steel martensite is shown as line A in Figure 1a, and the hatched area represents the same data with the recommended 1/3 to 1/2 correction factor.² Figure 1b gives a plot of the S-D effect as a percent of the uniaxial tensile yield strength, with the corrected values again represented by the hatched area.

The data are replotted in Figure 1b to see if the S-D effect is a constant percentage of the tensile yield strength since the tensile yield strength of the as-quenched martensite increases as the carbon is increased. Both parts of Figure 1 have been projected down to run through zero as is assumed by Hirth and Cohen. The hatched area of Figure 1a shows that the true S-D effect may range anywhere from about 5 to 15 percent for the 0.10 to 0.40 carbon range actually studied by Leslie and Sober.

For the 0.16 percent carbon content typical of HY-80 steel the S-D effect would be expected to be about 5 to 10 percent of the tensile yield strength of the as-quenched martensite, based on Figure 1b; the solid line in Figure 2a shows it to have been measured as 9 percent for an HY-80 steel austenitized at 1640F, water quenched and tempered at various temperatures.⁵

In addition, Figure 2a shows that tempering does not reduce the S-D effect to zero as the tempering temperature is increased. Instead, Figure 2a shows that the S-D effect increases to a maximum as the tempering temperature* is raised to about 600F corresponding to the precipitation of Fe₃C. It then decreases slightly as the temperature is raised to about 1150F, shows a slight inflection at about 1200F in the region where coherent alloy carbides start to form, and drops markedly from 10 to 5 percent as the temperature is raised closer to the lower critical temperature (A_{C1})** and as the alloy carbides grow and lose coherency. Beyond 1300F, slightly below the A_{C1}, there is a sudden and dramatic increase in the S-D effect from 5 to 28 percent as the tempering temperature passes into the critical temperature range where increasing proportions of austenite are formed and transform to fresh martensite upon quenching from the tempering temperature. This effect begins to reverse itself about 1400F, while the percent of austenite formed during tempering is still relatively small. This effect may be due either to the lowering of the carbon content of the ferrite matrix, accompanied by the change in

*One hour at temperature and then brine quenched from the tempering temperature.

**The A_{C1} is about 1330F for this steel.

shape and distribution of the hard particles in the soft matrix, or to the strain imposed on the ferrite lattice when the carbon-enriched austenite expands as it transforms to martensite. It is most likely the former, since examination of the lightly dashed line of Figure 2a for a coarse grained material originally austenitized at 2000F shows that the same effect appears to occur but at the temperatures normally associated with secondary hardening due to the formation of alloy carbides instead of at temperatures slightly above the A_{C1} . It is unfortunate that the coarse grained steel was not tempered in the range from 1330 to 1400F to see if a second peak occurred in the S-D effect, as the A_{C1} was exceeded during tempering of the coarse grained material.

Comparison of the solid and dashed lines on Figure 2a shows three things: (1) the S-D effect definitely does not decrease to zero as the tempering temperature is raised, but that it follows a closed loop (see Figure 2b); (2) the prior austenitic grain size of the steel can have a marked effect on the magnitude of the S-D effect; and (3) either secondary hardening or tempering above the lower critical temperature can have a marked effect on the S-D effect.

Figure 3 shows the effects of microstructure on the S-D effect, Figure 3a as a function of tempering temperature and Figure 3b as a function of strength.⁴ This material was austenitized at 2000F to produce a coarse grained austenitic structure prior to transformation. The open circles represent 100 percent martensite, the open triangles represent 3 percent bainite – 97 percent martensite, the closed triangles 17 percent bainite – 83 percent martensite, and the open squares 9 percent ferrite – 91 percent martensite. These figures show the following for tempered steels with a mixed microstructure: (1) the S-D effect persists at a significant level even after tempering; (2) it does not go to zero as the tempering temperature is raised; (3) disregarding the secondary hardening of the fully martensitic structure, the S-D effect is approximately the same for martensitic and semi-bainitic structures; (4) a semi-ferritic structure tends to have a markedly larger S-D effect at low tempering temperatures, up to about 1000F; and (5) after an 1150F temper for 1 hour, the percent S-D effect is quite consistent for all microstructures, 8 to 12 percent of the tensile yield strength.

The increased S-D effect for the semi-ferritic structure may be attributed to either the preferential lowering of the uniaxial tensile yield strength by the weak ferrite or to the increased carbon content of the martensite due to the carbon enrichment of the remaining austenite during the formation of the ferrite from austenite. The latter is not likely to be the significant factor, since the carbon increase would not be significant with the formation of 9 percent ferrite raising the carbon content of the remaining austenite from its original 0.16 percent to less than 0.18 percent. Examining Figure 1b, it can be seen that this is not enough to account for the increase in S-D effect from 5 to 10 percent observed for the semi-ferritic structure in Figure 9a.

Figure 9 is a plot of the percent S-D effect reported in Table 6 for 27 different steel plates heat treated to have a fully martensitic structure and then tempered at 1150F for 1 hour. The circles represent data on plates austenitized at 1640F for 1/2 hour to give a fine prior austenitic grain size, and the squares represent plates heated at 2000F for 1 hour to give a coarse prior austenitic grain size. The open circles represent plate material from heats made by one producer and the closed circles* represent heats made by the other. Figure 9 shows plots of the percent S-D effect as a function of carbon content, prior austenitic grain size, and ultimate tensile strength.

Figure 9 shows: (1) for the fine grained material where comparison can be made between producers, both producers' steel has the same magnitude S-D effect; (2) carbon content ranging from 0.14 to 0.22 percent had

*None of these heats was austenitized at 2000F.

no effect on the magnitude of the S-D effect; (3) prior austenitic grain size appears to have little or no effect on the magnitude of the S-D effect, considering the range of observed values for a given grain size; and (4) there appears to be a slight correlation between the magnitude of the S-D effect and the ultimate tensile strength of the steel. As can be seen from Figure 9c, the solid line representing the relationship between ultimate tensile strength and percent S-D effect is given by the equation below and indicates that the S-D effect (Y)

$$Y = \frac{1}{4} X - 23 \frac{3}{4} \quad (1)$$

goes to zero at an ultimate strength (X) of 95 ksi; it should be kept in mind that this projection is only for a fine grained, fully quenched martensitic HY-80 steel tempered for 1 hour at 1150F and that the usefulness of the equation should be limited to the range of the ultimate tensile strength data used to derive it, 117.5 to 138.8 psi. Equation (1) does not hold for other tempering temperatures, as can be seen from Figures 2 and 3, and the data in Figure 9 does not agree with the hypothesis that the S-D effect goes to zero at zero strength as is represented by the dashed line in Figure 9c.

Figure 10 is a plot of selected data from Figure 9; the number next to each data point plotted on Figure 11 refers to the plate identification number. Figure 10 gives only the data for the six plates which were tested after both the fine and coarse grain austenitizing treatments prior to quenching and tempering. Again, the S-D effect seems unrelated to carbon content, at least for the limited range shown, 0.16 to 0.22 percent C. There does appear to be a slight correlation of percent S-D effect to prior austenitic grain size and to ultimate strength. It should be kept in mind, however, that strength and grain size are not truly independent variables, since both yield strength and ultimate strength are a function of grain size,⁹ with the finer grained material expected and observed here to have higher strengths.

A comparison of the corrected data for 4300 steels in Figure 1 and for the 0.16 percent C HY-80 steel in the other figures shows that the S-D effect is almost a constant percentage of the tensile yield strength regardless of carbon content and regardless of the tempering temperature, as long as some secondary effect does not intervene, such as secondary hardening or exceeding the A_{C1} . This tends to support the contention that the S-D effect is the result of either substitutional elements straining the lattice, since their effect would exist even after tempering, or that the sign (\pm) of the normal stress across the glide plane during plastic deformation has a small but finite effect on the Peierls stress. The Lessel data for 1035 and 1045 steel show a similar effect for plain carbon steels and thus supports the latter.³

CONCLUSIONS

The following conclusions are based on the results of a study on the effects on the strength-differential effect of (1) commercial variation in composition and inclusion content, (2) variation in microstructure, prior austenitic grain size and microconstituents produced by quenching and by partial isothermal transformation at 1200 and 875F to produce ferrite and bainite prior to quenching, and (3) variation in strength level after an 1150F temper for one hour followed by a water quench to prevent embrittlement while cooling from the tempering temperature.

The HY-80 steel used in these studies is a high toughness, low impact transition temperature, quenched and tempered Ni-Cr-Mo steel of the type made to military specification MIL-S-16216 and ASTM Standard A 543-65. Sixteen coupons were obtained from one producer and 11 coupons from another. The average percent composition of the steels used in this study is given below:

0.18 C	0.10 Cu
0.36 MN	0.032 Acid Soluble Al
0.016 P	0.040 Total Al
0.019 S	0.002 Ti
0.30 Si	0.014 O ₂
2.75 Ni	0.007 N ₂
1.51 Cr	0.0001 H ₂
0.38 Mo	

The experimental data presented here and in References 4 and 5 show that the strength-differential effect is sensitive to microstructure and that it appears to be a function of the tensile yield strength rather than a function of the lattice strain caused by carbon atoms in the martensitic lattice as proposed by Hirth and Cohen in Reference 1. The following conclusions may be drawn from these data:

1. The data presented here on low-carbon Ni-Cr-Mo steels show that for the as-quenched structure, the magnitude of the S-D effect is markedly influenced by the prior austenitic grain size, which should not be the case if it were solely due to nonlinear strains induced in the body centered tetragonal lattice by interstitial carbon atoms.
2. The percent S-D effect observed in these studies persists and is essentially undiminished by tempering below the lower critical (A_{C1}).
3. The S-D effect observed in this study is about 8 percent (± 4 percent) of the tensile yield strength, regardless of strength level and carbon content for the group of low-carbon Ni-Cr-Mo steels having about the same composition and given the same tempering treatment.
4. The existence of the S-D effect in plain carbon steels, such as 1035 and 1045 reported by Lessels,³ indicates that large percentages of substitutional atoms distorting the iron lattice may not be necessary to produce the S-D effect.
5. Evaluation of Figure 1 in light of the data in Figures 2, 3, 9, and 10 indicates that the projection, $\Delta\sigma_{S-D}$ goes to zero as the carbon content goes to zero, may be correct, but this is attributed to the fact that $\Delta\sigma_{S-D}$ is effectively a constant percentage of the tensile yield strength which is also decreasing as the carbon content is lowered.
6. Based on the data reported in Reference 4, the S-D effect seems to be greatest for a mixed microstructure of ferrite and tempered martensite rather than a maximum for either a tempered fully martensitic or bainitic structure, and that this is probably an additive effect due in part to the more detrimental effect of small amounts of the weak ferrite in the stronger martensite when testing in tension as opposed to compression.
7. The relatively constant level of the S-D effect when expressed as a percent of the yield strength for various microstructures, as-quenched and as-quenched tempered below the lower critical temperature, suggests that interstitial carbon atoms are not the sole cause and probably not the major cause of the S-D effect.

8. The data presented here and in References 4 and 5 show that the percent S-D effect is markedly structure sensitive to such things as (a) prior austenitic grain size, (b) microstructural constituents, (c) tempering temperature, (d) type and distribution of carbides formed during tempering, and (e) tempering slightly above the lower critical (A_{C1}) temperature.

9. The results of the tests reported here on a low-carbon Ni-Cr-Mo steel treated to produce various microstructures strongly suggest that the S-D effect has a minimum value of about 5 percent of the tensile yield strength of the material in residual stress free, relatively isotropic, homogeneous structures not subject to unusual phenomenon such as preferring to twin to accomplish plastic deformation under compressive loading, and may range to much higher values due to its sensitivity to final microstructure.

10. The microstructural sensitivity of the percent S-D effect strongly indicates that the applied stress required to move a dislocation (the Peierls Stress) is being affected and is responsible for the S-D effect, and that, like frictional forces, it is sensitive to the sign of the resolved normal stresses across the glide plane, as well as to the apparent "roughness" of the glide plane.

ACKNOWLEDGMENT

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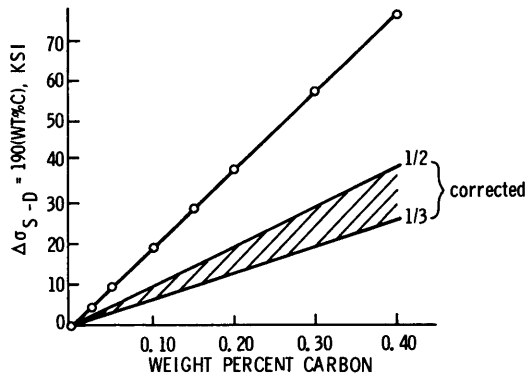


Figure 1a – Difference in Uniaxial Tensile and Compressive Yield Strength

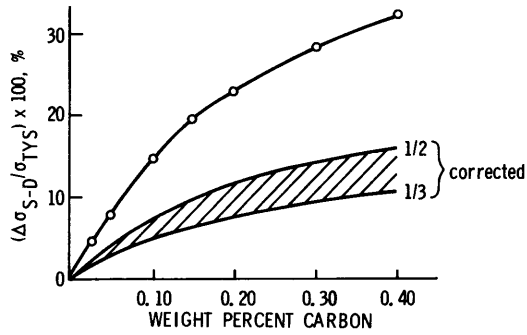


Figure 1b – Strength Differential as a Percent of Tensile Yield Strength

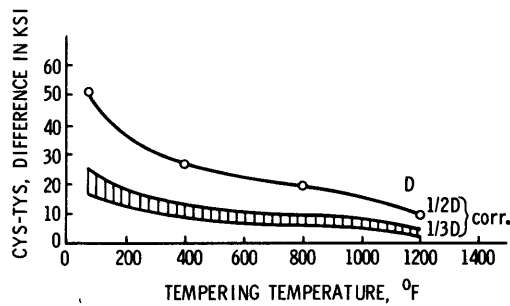
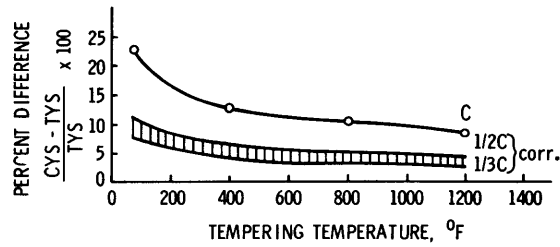


Figure 1c – S-D Effect in a Quenched and Tempered 4330 Steel

Figure 1 – Plot of the S-D Effect Observed by Leslie and Sober as a Function of the Carbon Content of As-Quenched, Fully Martensitic 4300 Series Steels and a Quenched and Tempered 4330 Steel

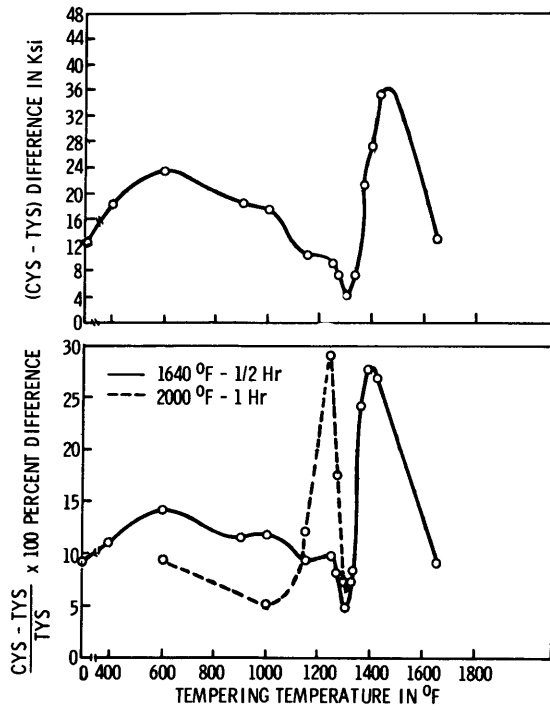


Figure 2a – Effects of Tempering Temperature on the S-D Effect

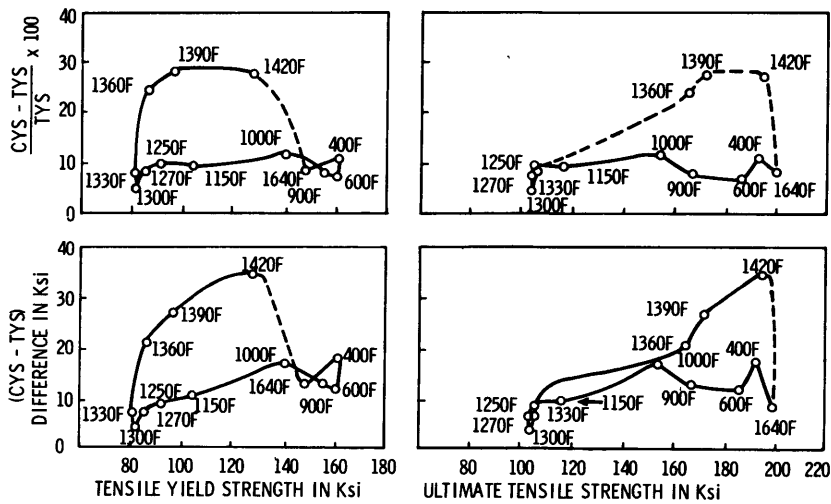


Figure 2b – Effects of Strength on the S-D Effect

Figure 2 – Effects of Tempering Temperature and Strength Level on the Relationship between Tensile and Compressive Yield Strength of an HY-80 Steel Austenitized at 1640 F for 1/2 Hour and Quenched to Produce a Fully Martensitic Structure

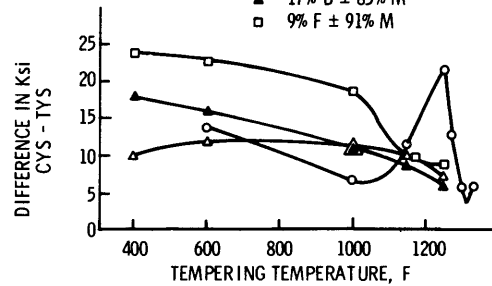
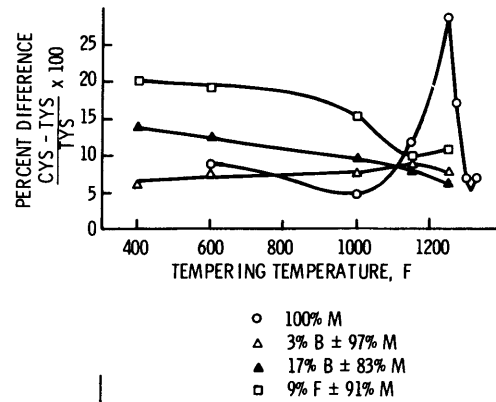


Figure 3a – Effects of Tempering Temperature and Microstructure on the S-D Effect

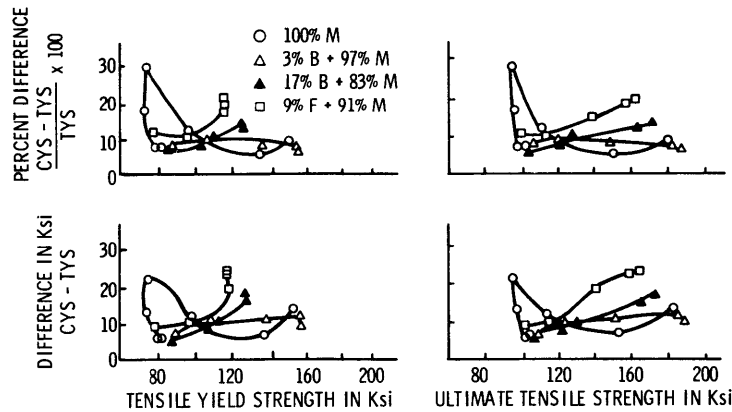


Figure 3b – Effects of Strength Level and Microstructure on the S-D Effect

Figure 3 – Effects of Tempering Temperature, Microstructure, and Strength Level on the S-D Effect Observed in an HY-80 Steel Austenitized at 2000 F for 1 Hour and Then Treated to Produce Various Microstructures

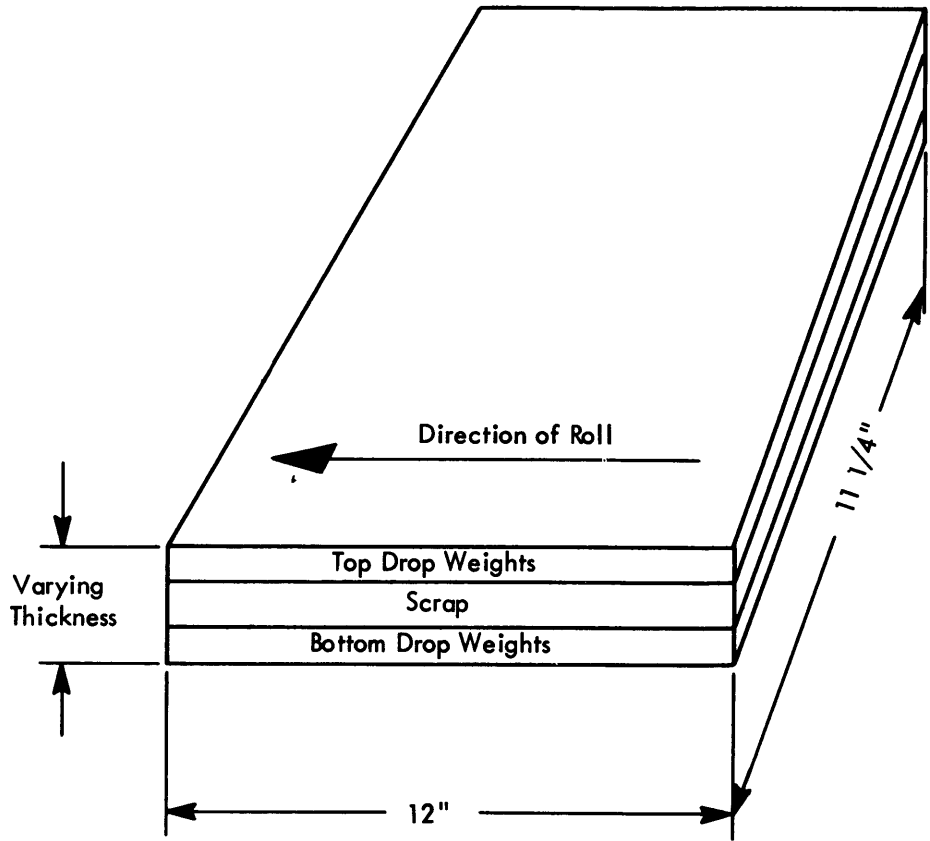
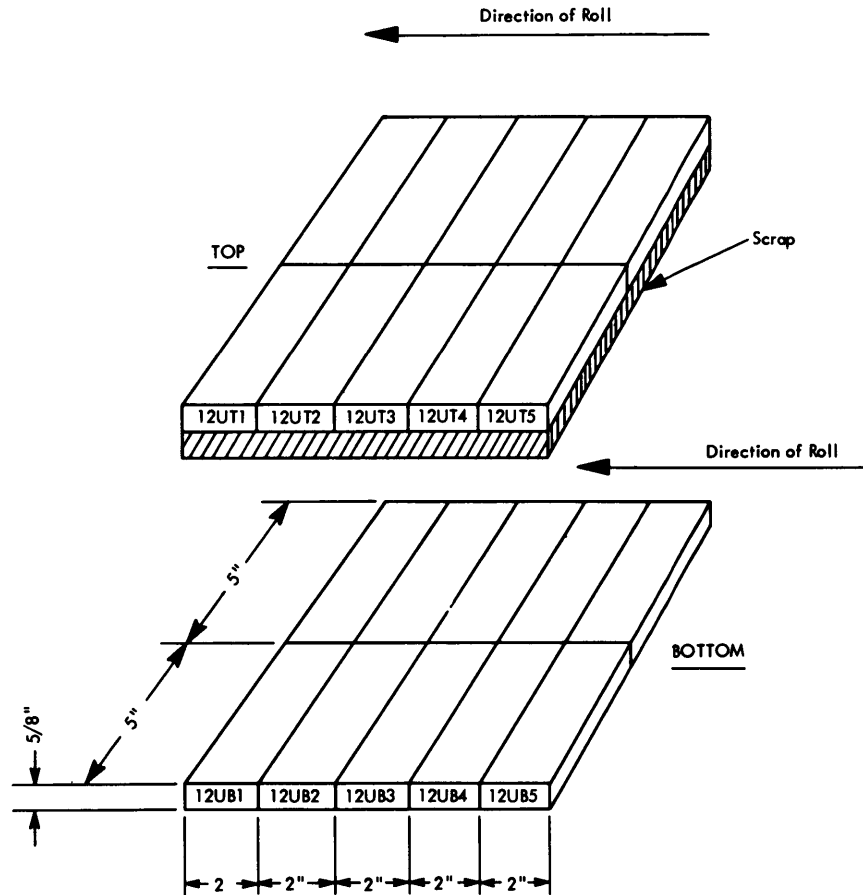
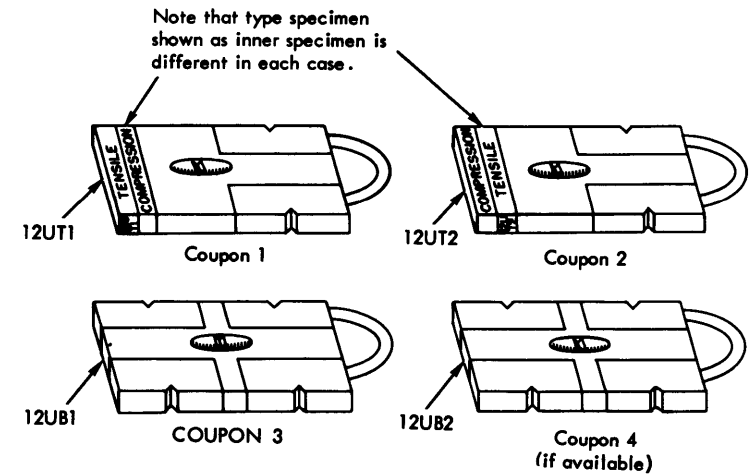


Figure 4 – Location of Material Used in This Study

**NOTES:**

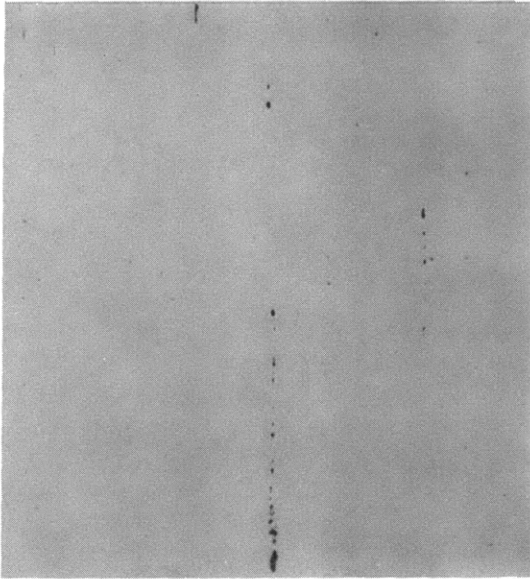
1. Scrap burned edges of coupon for at least 1/2 inch all around.
2. Stencil identification on 5/8" x 2 inch face shown on sketch. Identification should include: coupon number (12U, 13U, etc.), top (T) or bottom (B), and specimen number (1-10).

Figure 5 – Layout of Drop Weight Test Specimens



- NOTE: 1. Take the following specimens at the locations shown from each group of coupons:
- (A.) 2 longitudinal modified S-914 tensile specimens.
 - (B.) 2 longitudinal S-12914 compression specimens.
 - (C.) 10 transverse B-328-4 Charpy V-notch specimens. (See sheet No. 2 for additional instructions regarding the Charpy specimens.)
2. Identify each specimen with the following:
- (A.) Coupon No. (12UT4 for example)
 - (B.) Direction (L or T)
 - (C.) Specimen No. 1→4 (or 6 if necessary)
3. Identify tensile and Charpy specimens on both ends by stamping.
4. Identify compression specimens by scribing on the side only.
5. If any Charpy specimens have to come from under weld bead they should come from coupons 1 and 2.

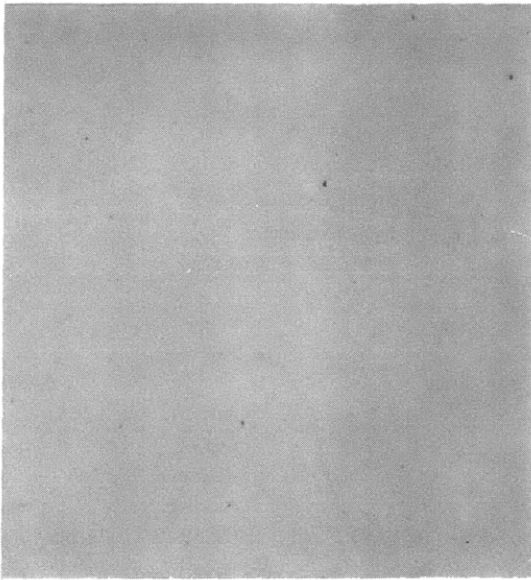
Figure 6 – Location of Mechanical Property and Transverse Charpy V-Notch Impact Specimens Made from Tested Drop Weight Specimens



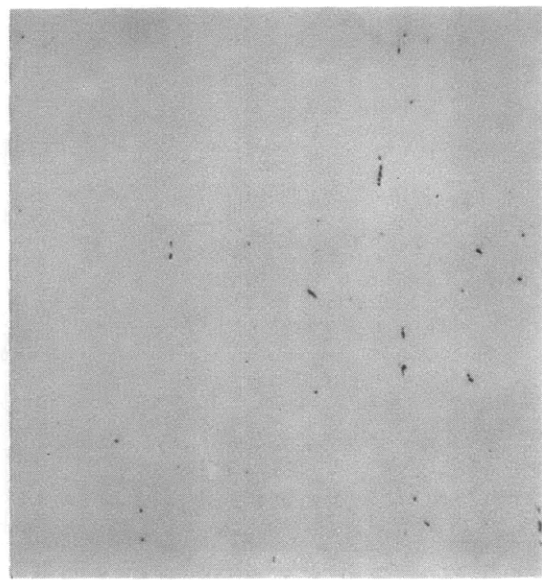
**Figure 7a – Mid-Thickness,
Longitudinal**



**Figure 7b – Mid-Thickness,
Transverse**



**Figure 7c – Adjacent to
Surface, Longitudinal**

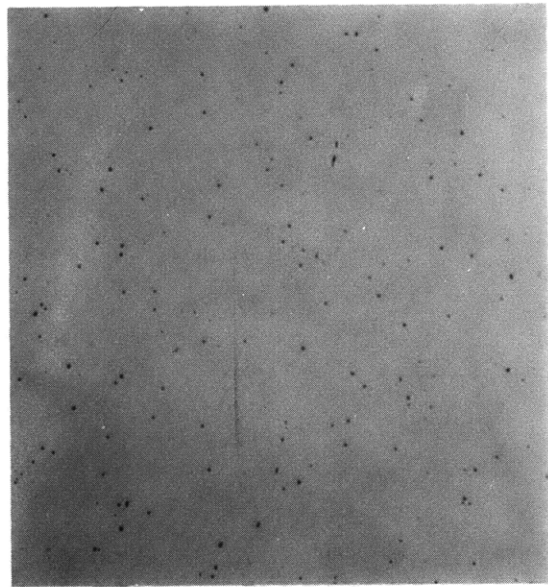


**Figure 7d – Adjacent to
Surface, Transverse**

Figure 7 – Photomicrographs of the Inclusion Content of Coupon X-18



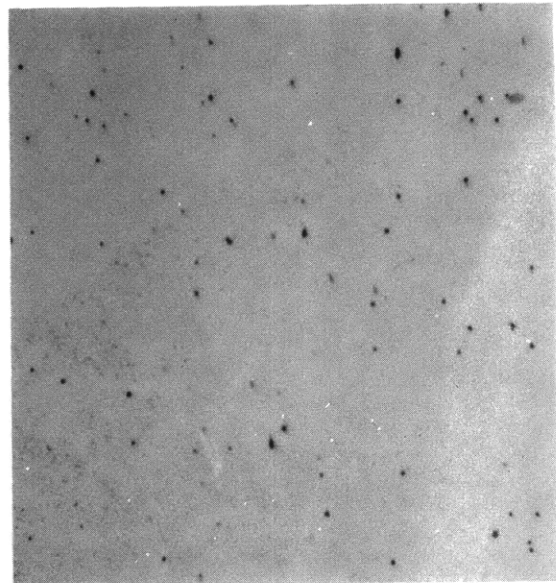
**Figure 8a – Mid-Thickness,
Longitudinal**



**Figure 8b – Mid-Thickness,
Transverse**



**Figure 8c – Adjacent to
Surface, Longitudinal**



**Figure 8d – Adjacent to
Surface, Transverse**

Figure 8 – Photomicrographs of the Inclusion Content of Coupon Y-1

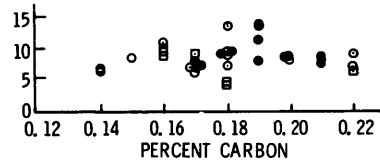


Figure 9a – Effect of Percent Carbon

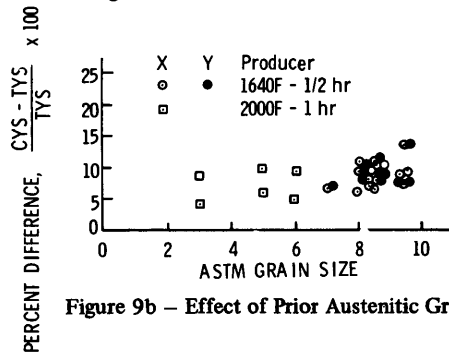


Figure 9b – Effect of Prior Austenitic Grain Size

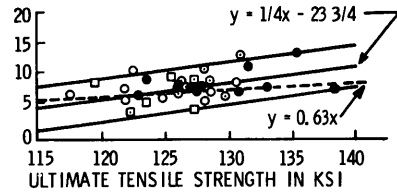


Figure 9c – Effect of the Ultimate Strength

Figure 9 – Effects of Percent Carbon, Prior Austenitic Grain Size, and Ultimate Tensile Strength on the S-D Effect Observed in 27 HY-80 Steel Plates Quenched to a Fully Martensitic Microstructure and Tempered at 1150 F for 1 Hour

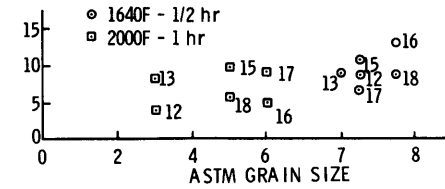


Figure 10a – Effect of Prior Austenitic Grain Size

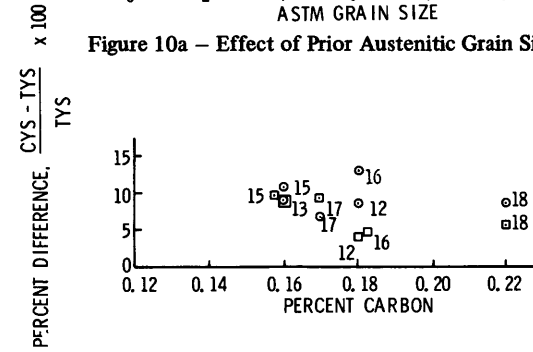


Figure 10b – Effect of Carbon Content

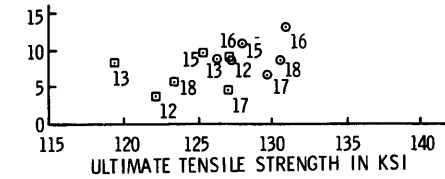


Figure 10c – Effect of Ultimate Tensile Strength

Figure 10 – Effects of Percent Carbon, Prior Austenitic Grain Size, and Ultimate Tensile Strength on the S-D Effect in Six HY-80 Steel Plates after Austenitizing at 1640 F for 1/2 Hour or 2000 F for 1 Hour, Quenching to a Fully Martensitic Structure, and Tempering at 1150 F for 1 Hour

TABLE 1

Producers' Reported Chemical Composition and Mechanical Properties

IDENT. NO.	HEAT NO.	LOC.	CHEMICAL COMPOSITION, PCT								COUPON SIZE		SPEC. DIR.	MECHANICAL PROPERTIES				LONGITUDINAL CHARTY V-NOTCH ENERGY, FT-LB	
			C	MN	P	S	CU	SI	NI	CR	MO	TXWXL, IN.		YIELD	TENSILE	ELONG. 2 IN., PCT	RED. IN AREA, PCT		
X- 1	65S399	L*	.16	.26	.011	.020		.23	2.60	1.35	.47								
		F	.16	.25	.014	.020		.22	2.59	1.40	.44	2X21X90	TRANS	88.1	108.0	24.0	63.8	101-108-108	
		R	.16	.28	.014	.019		.24	2.61	1.40	.36		LONG.	84.2	102.8	24.0	73.0	90- 89- 92	
X- 2	72S686	L	.14	.30	.013	.016		.24	2.85	1.53	.41								
		F	.15	.30	.015	.017		.22	2.83	1.58	.38	6X18X126	TRANS	80.8	100.7	26.0	70.1		
		R	.15	.28	.020	.017		.18	2.80	1.50	.39		LONG.	80.7	100.8	24.0	71.9	96- 96-104	
X- 3	72S686	L	.14	.30	.013	.016		.24	2.85	1.53	.41								
		F	.15	.30	.015	.017		.22	2.83	1.58	.38	1.62X18X118	TRANS	88.8	106.2	22.5	68.7	99- 97-97	
		R	.15	.28	.020	.017		.18	2.80	1.50	.39		LONG.	86.3	102.0	25.0	75.0	110- 85- 85	
X- 4	66S411	L	.16	.34	.016	.019		.28	2.84	1.66	.46								
		F	.19	.34	.016	.019		.27	2.85	1.72	.41	2X21X140	TRANS	93.7	113.7	23.0	57.2	84- 87- 86	
		R	.18	.34	.027	.020		.24	2.85	1.74	.40		LONG.	96.3	114.5	25.0	68.8	82- 92-102	
X- 5	66S357	L	.17	.30	.011	.018		.28	2.80	1.69	.45								
		F	.16	.32	.015	.019		.24	2.82	1.71	.42	2.13X17X133	TRANS	91.3	112.0	22.0	63.9	110-106-112	
		R	.16	.30	.022	.018		.22	2.83	1.71	.42		LONG.	89.1	110.4	24.0	71.5	110-110-124	
X- 6	65S799	L	.16	.32	.013	.015		.27	2.82	1.70	.46								
		F	.15	.30	.013	.015		.27	2.80	1.75	.45	3.75X18X76	TRANS	86.0	104.0	22.0	69.1		
		R	.14	.32	.012	.016		.25	2.85	1.75	.44		LONG.	80.1	101.2	22.0	74.3	124-130-120	
X- 8	66U040	L	.15	.25	.013	.017		.23	2.76	1.52	.35								
		F	.15	.26	.015	.016		.24	2.76	1.50	.27	2X23X140	TRANS	92.0	107.6	24.0	69.6	113-117-114	
		R	.14	.27	.021	.016		.18	2.74	1.38	.32		LONG.	87.5	105.7	24.0	63.9	121-125-113	
X- 9	70U032	L	.14	.29	.013	.016		.28	2.85	1.78	.38								
		F	.16	.28	.017	.017		.26	2.88	1.76	.36	1.5X24X100	TRANS	84.0	108.2	23.0	64.0	120-132-133	
		R	.14	.28	.012	.017		.25	2.81	1.77	.34		LONG.	80.1	102.1	24.0	75.2	127-117-124	
X-10	65U178	L	.16	.35	.018	.017		.27	3.00	1.62	.36								
		F	.16	.35	.012	.014		.24	2.91	1.68	.42	1.75X42X130	TRANS	90.8	111.2	22.0	64.4	123-125-122	
		R	.17	.38	.013	.015		.24	2.90	1.58	.36		LONG.	88.2	108.8	22.0	73.2	99- 91- 85	
X-11	X51294	L	.15	.26	.009	.011		.23	2.95	1.50	.50								
		F	.16	.28	.009	.012		.23	2.90	1.50	.48	2.13X36X139	TRANS	83.1	102.0	26.0	66.6	85- 82- 89	
		R	.17	.28	.010	.012		.23	2.89	1.49	.48		LONG.	80.1	104.9	26.0	71.6	99- 99- 94	
X-12	68U369	L	.16	.27	.012	.018		.27	3.00	1.58	.40								
		F	.17	.29	.011	.018		.22	2.99	1.50	.38	1.5X28X96	TRANS	91.6	109.2	23.0	66.8	127-135-131	
		R	.16	.29	.017	.017		.18	3.03	1.52	.37		LONG.	90.7	108.5	24.0	72.7	122-125-130	
X-13	68U369	L	.16	.27	.012	.018		.27	3.00	1.58	.40								
		F	.17	.29	.011	.018		.22	2.99	1.50	.38	4.5X23X96	TRANS	86.3	110.5	24.0	68.5		
		R	.16	.29	.017	.017		.18	3.03	1.52	.37		LONG.	82.6	103.6	24.0	73.8	115-102-119	

TABLE 1 (Continued)

X-18	66U397	L	.19	.27	.010	.017	.25	2.90	1.75	.41	2X20X135	TRANS LONG.	85.7 85.6	110.2 108.6	24.0 22.0	59.7 68.9	91- 68- 97 102-100- 70	
		F	.19	.29	.011	.016	.27	2.87	1.73	.39								
		R	.16	.25	.016	.017	.22	2.87	1.75	.40								
X-15	66U397	L	.19	.27	.010	.017	.25	2.90	1.75	.41	4.5X26X114	TRANS LONG.	92.8 89.7	110.8 109.1	20.0 24.0	67.9 62.9	94- 96- 97	
		F	.19	.29	.011	.016	.27	2.87	1.73	.39								
		R	.16	.25	.016	.017	.22	2.87	1.75	.40								
X-16	N51533	L	.16	.30	.014	.015	.22	2.92	1.63	.53	1.8X74X97	TRANS LONG.	91.1 89.8	109.3 108.9	22.0 24.0	63.5 68.3	95- 97- 97	
		F	.16	.31	.013	.014	.21	2.89	1.65	.51								
		R	.16	.33	.012	.015	.22	2.87	1.65	.52								
X-17	N51533	L	.16	.30	.014	.015	.22	2.92	1.63	.53	6X22X67	TRANS LONG.	91.2 88.2	108.5 106.8	25.0 26.0	68.3 71.4	86- 60- 65	
		F	.16	.31	.013	.014	.21	2.89	1.65	.51								
		R	.16	.33	.012	.015	.22	2.87	1.65	.52								
Y- 1	A2171-4A		.15	.22	.021	.024	.22	2.15	1.28	.27	.69X15X148	TRANS LONG.	87.4 90.5	104.5 106.5	26.0 26.0		90- 92- 97	
Y- 2	A2345-4			.26	.010	.017			1.28		2X15X161	TRANS LONG.	88.4 91.9	107.9 110.8	27.0 26.0		89- 96- 90	
Y- 3	A2472-10		.14	.19	.010	.026	.22	2.11	1.10	.26	.75X15X103	TRANS LONG.	86.2 82.8	103.4 101.0	23.0 25.0		96- 92- 95	
Y- 4	A2480-4C		.17	.35	.011	.022	.28	2.77	1.55	.37	2.62X15X75	TRANS LONG.	84.8 86.6	106.9 107.4	25.0 26.0		64- 71- 73	
Y- 5	A2848-2		.17	.31	.010	.023	.19	.26	2.81	1.49	.39	2.25X15X155	TRANS LONG.	88.5 87.1	109.0 107.9	24.0 26.0		87- 88- 86
Y- 6	20730-5		.17	.32	.010	.024	.17	.25	2.75	1.54	.42	2X15X121	TRANS LONG.	90.0 88.9	109.5 107.6	24.0 25.0		89- 90- 88
Y- 7	19540-6		.16	.35	.010	.025	.16	.30	2.74	1.67	.40	1.5X15X96	TRANS LONG.	85.1 82.1	104.1 102.5	25.0 26.0		86- 90- 90
Y- 8	19540-1		.16	.35	.010	.025	.16	.30	2.74	1.67	.40	2X15X161	TRANS LONG.	85.1 90.6	105.0 108.9	25.0 26.0		83- 81- 83
Y- 9	20844-2		.16	.29	.010	.025	.15	.25	2.76	1.47	.43	2X8.5X109	TRANS LONG.	88.6 88.4	108.4 108.1	25.0 28.0		109-109-108
Y-10	21222-2		.16	.34	.013	.023	.23	.27	2.80	1.53	.42	2.5X15X168	TRANS LONG.	85.1 90.6	105.0 108.9	25.0 26.0		78- 72- 75
Y-11	22616-1		.15	.25	.010	.025	.17	.20	2.85	1.42	.23	3.62X15X158	TRANS LONG.	81.6 80.1	103.5 99.0	25.0 27.0		45- 55- 53

* L IS LAOLE, F IS FRONT AND R IS REAR

TABLE 2

Actual Chemical Composition of Coupons

CHEMISTRY DATA

SPEC. NO.	AMT. C	AMT. MN	AMT. P	AMT. S	AMT. SI	AMT. NI	AMT. CR	AMT. MO	AMT. CU	ACID SOL AL	TOTAL AL	AMT. TI	AMT. V	AMT. CO	AMT. O(2)	AMT. N(2)	AMT H(2) PPM	D(I) 50M	GRAIN SIZE
X- 1	0.15	.280	.015	.018	0.210	2.590	1.450	0.360	0.076	.0280	.0400	.0040	.0010	.	.0110	.0050	1	5.78	8.50
X- 2	0.16	.380	.016	.021	0.300	2.750	1.470	0.350	0.048	.0320	.0360	.0010	.0020	.	.0050	.0060	0	7.83	8.00
X- 3	0.14	.390	.023	.020	0.270	2.760	1.470	0.350	0.049	.0230	.0260	.0010	.0020	.	.0150	.0030	1	7.95	7.00
X- 4	0.18	.360	.017	.018	0.260	2.930	1.800	0.310	0.075	.0240	.0420	.0050	.0010	.	.0490	.0080	7	9.33	9.50
X- 5	0.17	.300	.019	.017	0.230	2.780	1.740	0.320	0.044	.0230	.0470	.0050	.	.	.0480	.0060	3	8.96	8.50
X- 6	0.14	.290	.016	.011	0.210	2.710	1.730	0.290	0.040	.0230	.0330	.0040	.	.	.0200	.0070	1	5.71	8.00
X- 8	0.20	.370	.014	.023	0.320	2.750	1.400	0.300	0.047	.0410	.0460	.0010	.0020	.	.0150	.0090	1	8.41	8.50
X- 9	0.17	.420	.020	.015	0.300	2.810	1.620	0.310	0.048	.0320	.0360	.0010	.0030	.	.0060	.0080	0	8.85	8.50
X-10	0.22	.440	.015	.017	0.300	2.780	1.460	0.310	0.041	.0470	.0530	.0010	.0020	.	.0060	.0080	0	13.83	9.50
X-11	0.18	.390	.013	.015	0.310	2.890	1.420	0.470	0.074	.0400	.0480	.0020	.0030	.	.0130	.0110	0	10.27	8.50
X-12	0.18	.360	.021	.018	0.360	2.930	1.450	0.350	0.069	.0260	.0300	.0020	.0050	.0210	.0130	.0050	0	7.62	8.50
X-13	0.16	.340	.020	.017	0.320	2.850	1.420	0.350	0.054	.0360	.0410	.0010	.0040	.0190	.0110	.0060	0	7.58	8.00
X-18	0.22	.360	.017	.018	0.360	2.750	1.600	0.370	0.049	.0350	.0410	.0010	.0020	.0100	.0120	.0060	1	8.00	9.50
X-15	0.16	.370	.019	.018	0.300	2.700	1.610	0.350	0.060	.0360	.0410	.0010	.0040	.0110	.0100	.0050	0	7.53	8.50
X-16	0.18	.400	.017	.018	0.300	2.830	1.550	0.500	0.088	.0580	.0650	.0020	.0060	.0100	.0100	.0070	0	12.10	9.50
X-17	0.17	.390	.019	.022	0.310	2.780	1.540	0.500	0.101	.0640	.0710	.0020	.0060	.0100	.0080	.0070	1	10.50	8.50

SPEC. NO.	AMT. C	AMT. MN	AMT. P	AMT. S	AMT. SI	AMT. NI	AMT. CR	AMT. MO	AMT. CU	ACID SOL AL	TOTAL AL	AMT. TI	AMT. V	AMT. CO	AMT. O(2)	AMT. N(2)	AMT H(2) PPM	D(I) 50M	GRAIN SIZE
Y- 1	0.21	.320	.013	.023	0.310	2.150	1.200	0.270	0.160	.0220	.0250	.0010	.0020	.	.0040	.0090	2	4.68	8.50
Y- 2	0.19	.260	.008	.017	0.260	2.690	1.560	0.420	0.170	.0060	.0280	.0040	.	.	.0090	.0090	1	7.17	9.50
Y- 3	0.18	.290	.015	.021	0.320	2.100	1.100	0.270	0.170	.0320	.0360	.0010	.0010	.	.0160	.0090	1	3.87	8.50
Y- 4	0.21	.370	.010	.017	0.290	2.910	1.580	0.510	0.190	.0030	.0250	.0050	.	.	.0240	.0090	12	13.29	9.50
Y- 5	0.19	.320	.007	.019	0.280	2.890	1.520	0.340	0.260	.0150	.0360	.0040	.	.	.0310	.0100	1	8.14	8.50
Y- 6	0.18	.340	.011	.021	0.240	2.770	1.570	0.350	0.250	.0270	.0370	.0040	.	.	.0240	.0070	7	8.61	9.50
Y- 7	0.20	.380	.009	.025	0.300	2.750	1.670	0.550	0.190	.0270	.0470	.0050	.0010	.	.0250	.0090	1	13.47	8.50
Y- 8	0.17	.490	.015	.028	0.330	2.650	1.520	0.370	0.150	.0380	.0430	.0020	.0030	.	.0040	.0090	1	8.28	8.50
Y- 9	0.19	.420	.015	.018	0.300	2.730	1.400	0.400	0.150	.0320	.0360	.0010	.0020	.	.0040	.0070	0	10.02	8.50
Y-10	0.20	.490	.021	.018	0.330	2.700	1.450	0.380	0.190	.0380	.0420	.0110	.0020	.	.0110	.0070	2	9.69	8.50
Y-11	0.17	.380	.020	.022	0.290	2.790	1.400	0.390	0.160	.0100	.0130	.0010	.0030	.	.0080	.0020	0	9.89	7.00

TABLE 3

Inclusion Content

INCLUSION RATING AT THE SURFACE AND MID-THICKNESS
(ASTM E45-63, METHOD B)

SPECIMEN NUMBER	MID-THICKNESS						BOTTOM			SURFACE		
	LONGITUDINAL INCLUSIONS		BACK-GROUND	TRANSVERSE INCLUSIONS		BACK-GROUND	LONGITUDINAL INCLUSIONS		BACK-GROUND	TRANSVERSE INCLUSIONS		BACK-GROUND
	LONGEST	OTHERS		LONGEST	OTHERS		LONGEST	OTHERS		LONGEST	OTHERS	
X- 1	3.5 ^g	1.2 ²	A	1.0 ^{vd}	1.2 ¹	A	3.0 ^{vd}	1.6 ³	B	3.7 ^d	1.1 ⁴	A
X- 2	3.5 ^d	.0	D	2.5 ^d	1.0 ¹	C	1.5 ^{vd}	1.4 [†]	A	2.0 ^d	1.0 ¹	B
X- 3	5.0 ^d	2.1 ⁵	A	3.5 ^{vd}	1.5 ¹	A	3.5 ^g	1.5 ⁴	C	5.7 ^{vd}	1.7 ¹	B
X- 4	3.0 ^d	1.8 ²	A	1.5 ^d	1.3 ³	C	1.9 ^{vd}	1.2 ³	B	2.0 ^{vd}	.0	B
X- 5	8.0 ^g	1.6 ²	A	1.3 ^{vd}	.0	B	1.0 ^d	.0	A	6.0 ^d	3.6 ²	C
X- 6	6.1 ^g	1.9 ²	A	6.0 ^g	1.9 ⁶	B	2.8 ^d	2.0 ¹	A	1.5 ^{vd}	1.0 ¹	A
X- 8	3.5 ^g	2.1 ⁷	B	2.7 ^d	1.5 ⁴	A	3.5 ^d	1.7	D	4.2 ^{vd}	1.5 ³	D
X- 9	2.5 ^d	1.3 ⁷	B	3.5 ^g	1.0 ²	D	3.7 ^g	1.7 ⁴	C	.6 ^d	.0	A
X-10	2.3 ^d	1.7 ⁶	D	2.0 ^d	1.5 ³	B	1.5 ^d	1.1 ⁵	C	.7 ^{vd}	.0	A
X-11	2.8 ^g	1.4 ¹³	C	5.0 ^{vd}	1.5 ⁹	C	1.6 ^d	1.3 ¹⁰	D	1.3 ^d	.0	B
X-12	2.5 ^g	1.6 ¹³	C	1.0 ^{vd}	.0	A	4.3 ^{vd}	1.6 ²¹	C	2.7 ^{vd}	1.1	B
X-13	1.7 ^g	1.4 ¹¹	B	1.3 ^d	1.1 ²	A	1.1 ^d	1.0 ¹	D	2.2 ^g	1.2 ²	A
X-18	4.3 ^g	2.0 ²⁴	C	1.9 ^g	1.0 ³	C	3.2 ^{vd}	1.6 ¹⁶	C	1.2 ^d	.0	A
X-15	2.1 ^d	1.2 ⁸	D	1.5 ^g	.0	B	2.7 ^{vd}	1.3 ⁶	C	3.2 ^d	1.4 ¹	B
X-16	5.6 ^d	2.1 ⁶³	C	2.2 ^{vd}	1.5 ³	B	6.0 ^d	2.1 ²⁹	C	2.6 ^g	1.2 ³	D
X-17	4.3 ^g	1.6 ⁵	B	1.2 ^g	1.1 ³	A	2.1 ^d	1.3 ⁶	A	.6 ^d	.0	A
Y- 1	2.9 ^g	2.1 ⁵	C	7.4 ^d	2.7 ⁸	D	3.0 ^{vd}	.0	A	6.0 ^{vd}	.0	D
Y- 2	1.2 ^d	.0	A	4.0 ^d	1.1 ¹	A	1.9 ^d	1.2 ⁵	A	2.6 ^g	1.5 ¹	A
Y- 3	4.1 ^g	1.5 ⁷	A	7.0 ^d	.0	A	4.2 ^{vd}	2.6 ²	A	1.5 [†]	.0	A
Y- 4	2.4 ^g	1.2 ²	A	1.5 ^g	1.3 ¹	B	4.0 ^d	1.0 ¹	A	5.4 ^d	1.3 ³	A
Y- 5	4.2 ^d	1.3 ¹	A	5.5 ^{vd}	2.2 ⁴	C	4.0 ^{vd}	1.3 ⁴	A	2.5 ^{vd}	1.3 ²	C
Y- 6	2.5 ^d	1.1 ³	B	3.2 ^g	1.4 ⁴	C	2.0 ^{vd}	1.2 ⁴	B	1.1 ^{vd}	1.1 ¹	B
Y- 7	3.0 ^g	1.6 ¹³	C	1.0 ^d	.0	A	2.1 ^d	1.2 ¹	B	1.8	1.1 ²	A
Y- 8	2.0 ^d	1.0 ¹	B	1.2 ^d	1.1 ⁵	A	1.9 ^d	.0	A	2.0 ^d	1.0 ¹	D
Y- 9	2.1 ^d	1.4 ⁴	A	1.2 ^d	1.1 ¹	C	5.0 ^g	2.9 ³	A	2.0 ^{vd}	.0	B
Y-10	3.0 ^g	1.6 ¹⁶	C	3.0 ^d	1.3 ⁸	B	1.5 ^d	1.3 ¹	C	3.1 ^d	1.7 ²	A
Y-11	2.5 ^d	1.7 ⁴	A	3.0 ^{vd}	1.0 ¹	A	2.5 ^d	.0	B	.7 ^d	.0	A

TABLE 4

Prior Austenitic Grain Size

SPECIMEN NUMBER	AS RECEIVED MATERIAL MID-THICKNESS	AUSTENITIZING TREATMENT *	
		1640 F-1/2 HR-W.Q.	2000 F-1 HR-W.Q.
X- 1	6	8.5	
X- 2	6	8	
X- 3	5	7	
X- 4	6	9.5	
X- 5	4	8.5	
X- 6	7	8.	
X- 8	5	8.5	
X- 9	6	8.5	
X-10	4	9.5	
X-11	6	8.5	
X-12	8	8.5	3
X-13	6	8	3
X-18	5	9.5	5
X-15	7	8.5	5
X-16	5	9.5	6
X-17	6	8.5	6
Y- 1	7	8.5	
Y- 2	7	9.5	
Y- 3	7	8.5	
Y- 4	5	9.5	
Y- 5	6	8.5	
Y- 6	5	9.5	
Y- 7	5	8.5	
Y- 8	7	8.5	
Y- 9	5	8.5	
Y-10	7	8.5	
Y-11	4	7	

* AFTER AUSTENITIZING AND QUENCHING, -120 F 1 HR,
TEMPERED AT 1150 F 1 HR THEN WATER QUENCHED.

TABLE 5

Percent Microconstituents Produced by Various Isothermal Treatments

AUSTENITIZING

TEMPERATURE TIME	1640 F 1/2 HR	1640 F 1/2 HR	1640 F 1/2 HR	1640 F 1/2 HR	1640 F 1/2 HR	2000 F 1 HR	
ISOTHERMAL TEMPERATURE TIME	875 F 152 SEC	875 F 1600 SEC	1200 F 3350 SEC	1200 F 8500 SEC	1200 F 8500 SEC	875 F 1600 SEC	
MICRO- CONSTITUENTS	BAINITE	BAINITE	FERRITE	PEARLITE	FERRITE	PEARLITE	BAINITE
COUPON NUMBER							

X- 1	29	54	31	0	67	11	
X- 2	14	54	TRACE	0	2	0	
X- 3	21		31	0			
X- 4	2		11	0			
X- 5	1		0	0			
X- 6	0		1	0			
X- 8	14		5	0			
X- 9	7		26	0			
X-10	1		TRACE	0			
X-11	0		0	0			
X-12	1	50	TRACE	0	1	0	63
X-13	14	60	TRACE	0	TRACE	0	67
X-18	TRACE	21	0	0	0	0	56
X-15	1	27	0	0	0	0	58
X-16	0	16	TRACE	0	TRACE	0	58
X-17	0	32	0	0	TRACE	0	30
Y- 1	53		50	32			
Y- 2	9	41	22	0	53	4	
Y- 3	76		69	23			
Y- 4	1	39	7	0	21	0	
Y- 5	3		17	0			
Y- 6	1		12	0			
Y- 7	1		3	0			
Y- 8	1		1	0			
Y- 9	6		2	0			
Y-10	4		1	0			
Y-11	27	63	3	0	44	0	

TABLE 6

Results of Mechanical Property Tests

Table 6a – Specimens Austenitized at 1640 F for 1/2 Hour, Water Quenched and Tempered at 1150 F for 1 Hour

SPEC. NO.	PCT. BANITE	PCT. FERRITE	PCT. PEARLITE	MARTEN-SITE	ISO-THERMAL HOLDING TEMP. (F)	ISO-THERMAL HOLDING TIME (SEC)	ULT. TEN. STH. (PSI)	TEN. YIELD STH. (PSI)	PCT. EL. IN.	PCT. RED. IN. AREA	COMP. YIELD STH. (PSI)	NDT TEMP (F)
X- 1	0.	0.	0.	100.0	0	0	125907	113820	22.0	73.9	123334	-190
X- 2	0.	0.	0.	100.0	0	0	122392	107343	24.0	73.5	118860	-170
X- 3	0.	0.	0.	100.0	0	0	124850	114128	22.0	72.7	121686	-120
X- 4	0.	0.	0.	100.0	0	0	128526	113023	23.0	72.6	123578	-190
X- 5	0.	0.	0.	100.0	0	0	128115	113043	22.0	72.5	119830	-190
X- 6	0.	0.	0.	100.0	0	0	121792	107759	23.0	74.5	114383	-170
X- 8	0.	0.	0.	100.0	0	0	121756	109280	23.0	74.7	118068	-190
X- 9	0.	0.	0.	100.0	0	0	117448	104138	23.0	76.5	111430	-190
X-10	0.	0.	0.	100.0	0	0	125926	111361	23.0	74.2	119220	-190
X-11	0.	0.	0.	100.0	0	0	128527	118118	22.0	69.6	126668	-190
X-12	0.	0.	0.	100.0	0	0	127346	114571	22.5	74.7	124813	-180
X-13	0.	0.	0.	100.0	0	0	126275	112988	23.0	72.7	123228	-180
X-18	0.	0.	0.	100.0	0	0	130601	116000	23.0	72.8	126265	-190
X-15	0.	0.	0.	100.0	0	0	128074	112289	23.0	73.4	124506	-190
X-16	0.	0.	0.	100.0	0	0	130870	116883	23.0	74.5	132264	-210
X-17	0.	0.	0.	100.0	0	0	129743	116617	22.0	71.9	124533	-180
Y- 1	0.	0.	0.	100.0	0	0	125874	113686	21.0	70.2	122785	-210
Y- 2	0.	0.	0.	100.0	0	0	135325	122294	22.0	70.3	138652	-190
Y- 3	0.	0.	0.	100.0	0	0	123600	114600	22.0	72.0	125312	-230
Y- 4	0.	0.	0.	100.0	0	0	138777	124698	22.0	58.7	134012	-210
Y- 5	0.	0.	0.	100.0	0	0	131421	115871	22.0	69.1	129079	-230
Y- 6	0.	0.	0.	100.0	0	0	130814	115277	23.0	71.7	123819	-190
Y- 7	0.	0.	0.	100.0	0	0	133000	117150	21.0	66.5	126756	-220
Y- 8	0.	0.	0.	100.0	0	0	127600	112100	22.0	71.8	120685	-210
Y- 9	0.	0.	0.	100.0	0	0	126527	111611	22.0	74.2	120598	-190
Y-10	0.	0.	0.	100.0	0	0	128001	112000	23.0	71.1	121602	-230
Y-11	0.	0.	0.	100.0	0	0	122955	109532	22.5	71.3	117168	-150

Table 6b – Specimens Austenitized at 1640 F for 1/2 Hour, Isothermally Treated at 875 F for 152 Seconds, Water Quenched and Tempered at 1150 F for 1 Hour

SPEC. NO.	PCT. BANITE	PCT. FERRITE	PCT. PEARLITE	MARTEN-SITE	ISO-THERMAL HOLDING TEMP. (F)	ISO-THERMAL HOLDING TIME (SEC)	ULT. TEN. STH. (PSI)	TEN. YIELD STH. (PSI)	PCT. EL. IN.	PCT. RED. IN. AREA	COMP. YIELD STH. (PSI)	NDT TEMP (F)
X- 1	29.0	0.	0.	71.0	875	152	120984	105422	24.0	75.3	111571	-210
X- 2	14.2	0.	0.	85.8	875	152	119382	103331	24.0	4.3	111926	-150
X- 3	21.2	0.	0.	78.8	875	152	118600	105099	22.0	73.1	111957	-110
X- 4	1.5	0.	0.	98.5	875	152	127008	109036	23.0	72.9	118620	-191
X- 5	0.6	0.	0.	99.4	875	152	127204	109175	23.0	72.9	119614	-191
X- 6	0.3	0.	0.	99.7	875	152	121291	105253	24.0	75.8	115366	-191
X- 8	13.5	0.	0.	86.5	875	152	116168	101897	23.0	73.9	110132	-190
X- 9	6.8	0.	0.	93.2	875	152	118165	104594	23.0	76.1	113073	-180
X-10	1.1	0.	0.	98.9	875	152	125459	116784	23.0	76.0	118691	-210
X-11	0.	0.	0.	100.0	875	152	126658	113550	23.0	72.0	124016	-190
X-12	1.1	0.	0.	98.9	875	152	120801	105399	23.0	75.1	115540	-170
X-13	14.1	0.	0.	85.9	875	152	118327	102789	23.0	73.7	109078	-170
X-18	0.7	0.	0.	99.3	875	152	124799	107999	23.0	74.7	120065	-220
X-15	0.7	0.	0.	99.3	875	152	123355	106389	23.0	73.1	115127	-210
X-16	0.	0.	0.	100.0	875	152	123755	109473	23.0	74.3	119487	-210
X-17	0.	0.	0.	100.0	875	152	123307	108168	23.0	72.9	120074	-190
Y- 1	53.0	0.	0.	47.0	875	152	114885	98902	23.0	70.9	106646	-170
Y- 2	9.0	0.	0.	91.0	875	152	130814	116780	23.0	72.1	125320	-240
Y- 3	76.0	0.	0.	24.0	875	152	107480	92724	24.0	73.1	103014	-170
Y- 4	0.7	0.	0.	99.3	875	152	134932	119382	22.0	69.9	132787	-210
Y- 5	2.2	0.	0.	97.8	875	152	131526	116466	22.0	71.3	129051	-210
Y- 6	0.9	0.	0.	99.1	875	152	126906	109350	23.0	73.3	124078	-210
Y- 7	1.4	0.	0.	98.6	875	152	132600	116201	21.0	67.9	123508	-220
Y- 8	1.1	0.	0.	98.9	875	152	128614	113709	22.5	71.1	124195	-220
Y- 9	6.3	0.	0.	93.7	875	152	125896	109761	22.5	74.3	121646	-210
Y-10	4.0	0.	0.	96.0	875	152	129155	114828	22.0	71.6	123200	-210
Y-11	27.0	0.	0.	73.0	875	152	119841	106182	22.0	71.9	112789	-150

Table 6c – Specimens Austenitized at 1640 F for 1/2 Hour, Isothermally Treated at 875 F for 1600 Seconds, Water Quenched and Tempered at 1150 F for 1 Hour

SPEC. NO.	PCT. BANITE	PCT. FERRITE	PCT. PEARLITE	MARTEN-SITE	ISO-THERMAL HOLDING TEMP. (F)	ISO-THERMAL HOLDING TIME (SEC)	ULT. TEN. (PSI)	TEN. YIELD (PSI)	PCT. EL. IN.	PCT. RED. IN AREA	COMP. YIELD (PSI)	NDT TEMP (F)
X- 1	54.0	0.	0.	46.0	875	1600	112337	92530	24.0	73.4	104176	-130
X- 2	53.7	0.	0.	46.3	875	1600	109457	93914	25.0	74.2	104142	-170
X-12	49.4	0.	0.	50.6	875	1600	114795	98893	23.0	75.3	104176	-190
X-13	60.0	0.	0.	40.0	875	1600	115485	98102	23.0	73.2	105022	-190
X-18	20.7	0.	0.	79.3	875	1600	124573	108323	23.0	73.7	115765	-190
X-15	26.9	0.	0.	73.1	875	1600	123476	105394	22.0	72.5	113029	-190
X-16	16.2	0.	0.	83.8	875	1600	124000	110000	23.0	73.8	116458	-190
X-17	31.5	0.	0.	68.5	875	1600	121878	105794	22.0	73.3	112890	-170
Y- 2	40.8	0.	0.	59.2	875	1600	123818	107536	23.0	71.3	116268	-180
Y- 4	38.6	0.	0.	61.4	875	1600	124374	109128	23.0	70.5	118196	-190
Y-11	63.0	0.	0.	37.0	875	1600	112949	96912	23.0	71.1	101911	-150

Table 6d – Specimens Austenitized at 1640 F for 1/2 Hour, Isothermally Treated at 1200 F for 3350 Seconds, Water Quenched and Tempered at 1150 F for 1 Hour

SPEC. NO.	PCT. BANITE	PCT. FERRITE	PCT. PEARLITE	MARTEN-SITE	ISO-THERMAL HOLDING TEMP. (F)	ISO-THERMAL HOLDING TIME (SEC)	ULT. TEN. (PSI)	TEN. YIELD (PSI)	PCT. EL. IN.	PCT. RED. IN AREA	COMP. YIELD (PSI)	NDT TEMP (F)
X- 1	0.	31.0	0.	69.0	1200	3350	108000	91000	23.0	74.0	110597	-250
X- 2	0.	0.1	0.	99.9	1200	3350	117581	103248	24.0	74.9	110570	-170
X- 3	0.	31.3	0.	68.7	1200	3350	104296	81169	24.0	71.0	88185	-190
X- 4	0.	11.4	0.	88.6	1200	3350	118690	98588	24.0	74.2	109708	-210
X- 5	0.	0.	0.	100.0	1200	3350	124199	105168	23.0	74.2	116642	-190
X- 6	0.	0.5	0.	99.5	1200	3350	113545	979700	25.0	77.3	104710	-170
X- 8	0.	5.0	0.	95.0	1200	3350	117050	103446	22.5	75.5	111654	-190
X- 9	0.	26.1	0.	73.9	1200	3350	108566	88396	24.0	76.3	96054	-210
X-10	0.	0.1	0.	99.9	1200	3350	121592	106070	24.0	75.5	113179	-210
X-11	0.	0.	0.	100.0	1200	3350	122954	108184	23.0	71.3	115221	-230
X-12	0.	0.5	0.	99.5	1200	3350	122986	108459	23.0	75.3	119321	-190
X-13	0.	0.1	0.	99.9	1200	3350	121315	106574	23.0	73.5	117327	-190
X-18	0.	0.	0.	100.0	1200	3350	128016	112414	23.0	74.5	110597	-210
X-15	0.	0.	0.	100.0	1200	3350	126620	110320	23.0	72.1	119382	-190
X-16	0.	0.2	0.	99.8	1200	3350	127218	113260	23.0	73.9	124414	-210
X-17	0.	0.	0.	100.0	1200	3350	127019	113360	22.0	71.9	120455	-190
Y- 1	0.	50.2	32.1	17.7	1200	3350	88823	65869	29.0	62.5	71587	-100
Y- 2	0.	22.3	0.	77.7	1200	3350	129000	113000	23.0	73.0	110597	-250
Y- 3	0.	69.0	22.8	8.1	1200	3350	82672	61554	31.0	69.1	65068	-110
Y- 4	0.	6.5	0.	93.5	1200	3350	124398	107343	24.0	72.5	116627	-210
Y- 5	0.	17.0	0.	83.0	1200	3350	123395	108347	23.0	70.5	115424	-230
Y- 6	0.	12.3	0.	87.7	1200	3350	147000	115000	23.0	69.0	110597	-230
Y- 7	0.	3.3	0.	96.7	1200	3350	130609	113161	21.0	67.3	124597	-230
Y- 8	0.	0.5	0.	99.5	1200	3350	124852	109544	23.0	71.6	117527	-250
Y- 9	0.	1.8	0.	98.2	1200	3350	123200	107800	23.0	74.8	117051	-250
Y-10	0.	0.7	0.	99.3	1200	3350	124303	103765	23.0	71.5	116863	-230
Y-11	0.	2.6	0.	97.4	1200	3350	116485	100900	23.0	71.8	109661	-190

Table 6e – Specimens Austenitized at 1640 F for 1/2 Hour, Isothermally Treated at 1200 F for 8500 Seconds, Water Quenched and Tempered at 1150 F for 1 Hour

SPEC. NO.	PCT. BANITE	PCT. FERRITE	PCT. PEARLITE	MARTEN-SITE	ISO-THERMAL HOLDING TEMP. (F)	ISO-THERMAL HOLDING TIME (SEC)	ULT. TEN. STH. (PSI)	TEN. YIELD STH. (PSI)	PCT. EL. IN. 2	PCT. RED. IN. AREA	COMP. YIELD STH. (PSI)	NDT TEMP (F)
X- 1	0.	67.0	10.7	22.2	1200	8500	94518	68993	28.0	67.1	75343	-180
X- 2	0.	1.4	0.	98.6	1200	8500	115339	100598	24.0	74.3	106871	-190
X-12	0.	1.1	0.	98.9	1200	8500	120400	106200	23.0	74.8	114385	-190
X-13	0.	0.1	0.	99.9	1200	8500	120041	104188	23.0	73.7	111387	-170
X-18	0.	0.	0.	100.0	1200	8500	125001	108417	22.5	73.3	116716	-230
X-15	0.	0.	0.	100.0	1200	8500	124378	107428	23.0	72.7	113091	-190
X-16	0.	0.4	0.	99.6	1200	8500	124400	108598	22.0	73.6	117777	-220
X-17	0.	0.3	0.	99.7	1200	8500	123753	108892	23.0	72.1	114733	-190
Y- 2	0.	53.2	4.3	42.5	1200	8500	106878	81354	26.0	65.2	88944	-210
Y- 4	0.	20.5	0.	79.5	1200	8500	127019	108176	24.0	67.0	118509	-230
Y-11	0.	44.3	0.	55.7	1200	8500	103984	79931	25.0	70.3	85639	-210

Table 6f – Specimens Austenitized at 2000 F for 1 Hour, Transferred to 1640 F for 5 Minutes Minimum, Water Quenched and Tempered at 1150 F for 1 Hour

SPEC. NO.	PCT. BANITE	PCT. FERRITE	PCT. PEARLITE	MARTEN-SITE	ISO-THERMAL HOLDING TEMP. (F)	ISO-THERMAL HOLDING TIME (SEC)	ULT. TEN. STH. (PSI)	TEN. YIELD STH. (PSI)	PCT. EL. IN. 2	PCT. RED. IN. AREA	COMP. YIELD STH. (PSI)	NDT TEMP (F)
X-12	0.	0.	0.	100.0	0	0	122123	107608	21.5	72.2	111884	-110
X-13	0.	0.	0.	100.0	0	0	119341	104686	22.0	71.3	113546	-90
X-18	0.	0.	0.	100.0	0	0	123430	106780	22.0	72.3	113127	-100
X-15	0.	0.	0.	100.0	0	0	125423	108174	22.0	72.5	118627	-110
X-16	0.	0.	0.	100.0	0	0	127199	113600	21.5	71.6	118993	-110
X-17	0.	0.	0.	100.0	0	0	127146	112874	22.0	71.5	123237	-110

Table 6g – Specimens Austenitized at 2000 F for 1 Hour, Transferred to 1640 F for 5 Minutes Minimum, Isothermally Treated at 875 F for 1600 Seconds, Water Quenched and Tempered at 1150 F for 1 Hour

SPEC. NO.	PCT. BANITE	PCT. FERRITE	PCT. PEARLITE	MARTEN-SITE	ISO-THERMAL HOLDING TEMP. (F)	ISO-THERMAL HOLDING TIME (SEC)	ULT. TEN. STH. (PSI)	TEN. YIELD STH. (PSI)	PCT. EL. IN. 2	PCT. RED. IN. AREA	COMP. YIELD STH. (PSI)	NDT TEMP (F)
X-12	62.8	0.	0.	37.2	875	1600	116733	101494	22.0	71.3	107197	-80
X-13	66.9	0.	0.	33.1	875	1600	115547	100051	21.5	70.4	103241	-60
X-18	56.2	0.	0.	43.8	875	1600	121480	104797	22.0	70.7	108874	-80
X-15	58.4	0.	0.	41.6	875	1600	123000	106300	22.0	67.8	112219	-60
X-16	56.7	0.	0.	43.3	875	1600	124029	109772	22.0	71.1	116274	-110
X-17	29.9	0.	0.	70.1	875	1600	125000	111000	22.0	70.3	115943	-120

APPENDIX A

REAGENTS USED TO DEVELOP PRIOR AUSTENITIC GRAIN BOUNDARIES

ETHERAL PICRIC ACID PRIOR AUSTENITIC GRAIN BOUNDARY OR TEMPER BRITTLENESS ECHANT* AND ETCHING TECHNIQUE

The ethereal picric acid reagent consists of the following reagents:

Picric Acid (solid with 11 percent water)	50 grms
Purified Ethyl Ether	250 ml
Zephiran Chloride (12.8 percent solution)**	10 ml
Water	240 ml

The reagent is prepared and used as follows:

1. Dissolve the picric acid in ether (most of the picric acid will remain undissolved). Then add 250 ml of zephiran chloride and water solution.

2. Shake thoroughly; keep in tightly stoppered bottle. This mixture separates into two distinct layers. The top layer is saturated solution of picric acid in ether with some zephiran chloride. Bottom layer is aqueous zephiran chloride – picric acid solution.

3. Let stand overnight. No further shaking is necessary or desirable.

4. Decant small portion of the top layer and dilute with 1/3 volume ether. (Prevents supersaturation with picric acid.) If picric acid crystals precipitate upon the specimen, a badly pitted, uneven etch results. After use, the etching solution is usually returned to the bottle. That is, the etchant may be reused and like Picral appears to improve with use.

5. Technique:

a. To develop prior austenitic grain boundaries, immerse without agitation for 1 to 15 minutes. The exact time depends on how long it takes to develop the network of the prior austenitic grain boundaries. Keep solution at 80 ± 2 F. Repeat, if too lightly etched, after washing. If removed for cursory examination, wash with alcohol before re-immersion because of a precipitated layer of picric acid crystals. This film is insoluble in water.

b. To determine the presence of embrittling grain boundary precipitates the specimen is taken after step (a) and repolished to remove most of the grain boundary network. Repolish for approximately 1 minute on final wheel. Note: this polishing will take 6 minutes if a vibratory polisher is used. This etch is removed but grain boundary pitting due to an embrittling constituent present in the prior austenitic grain boundaries will persist and provide a definite indication in an improperly heat treated or temper embrittled specimen; a properly treated specimen will only have an occasional pit remaining at the location of the prior austenitic grain boundaries, whereas a network of pits is characteristic of a material suffering from temper embrittlement.

*Transactions of A.S.M., Vol. XXXIX (1947) pp. 109–138 inclusive.

**A brand of benzalkonium chloride manufactured by Winthrop-Stearns, New York, New York.

ALTERNATE GRAIN BOUNDARY ETCH

The picric acid and alcohol reagent consists of the following reagents:

Picric Acid (11 percent water)	Excess
Ethyl Alcohol (180 proof, undenatured)	100 ml
Sodium Tridecylbenzene Sulfonate*	1 gm

The reagent is prepared as follows:

1. Prepare a saturated solution of picric acid in ethyl alcohol and add the tridecylbenzene.
2. To etch, immerse for 3 to 6 seconds to develop network or prior austenitic grain boundaries.
3. Rinse in stream of hot water and then re-rinse immediately with 180 proof undenatured ethyl alcohol.
4. Repeat rinses in step (3) above.
5. Dry with hot air after second alcohol rinse.

*One source is Chemithon Corporation, 5430 W. Marginal Way, Seattle, Washington.

Note: If this etch is used instead of the ethereal picric acid etch, it should be used first, before the one percent Nital etch, when using the triple etching technique to identify isothermal transformation products. See Appendix B.

APPENDIX B

ETCHING TECHNIQUES FOR IDENTIFYING ISOTHERMAL PRODUCTS

REAGENTS

The first step is to prepare the following four solutions:

1. One Percent Nital.
2. Saturated Picric Acid Etch. (See Appendix A for details describing how to prepare this etch or its alternate.)
 - a. Before using the ethereal picric acid etch described in enclosure (2), add two drops of saturated Zephiran Chloride (a brand of benzalkonium chloride manufactured by Winthrop-Stearns, N. Y. 18, N. Y., and available in drug stores) to each 50 ml.
 - b. Appendix A gives an alternate alcohol base etch that will do an adequate job of revealing the prior austenitic grain boundaries. This alcohol base etch has the distinct advantage of ease of preparation and a much less objectionable odor.
3. Neutralizing Rinse. Add 10 ml concentrated ammonium hydroxide to 90 ml ethyl alcohol.
4. Twenty Percent Sodium Metabisulfite Solution.
 - a. Add 20 grams of Fisher's sodium metabisulfite (Cat. No. S-243) to 80 ml of distilled water. Allow to go completely into solution. Add two drops of saturated aerosol solution; aerosol is a wetting agent.
 - b. Filter to remove any particles before using.
 - c. Solution must be made fresh every day since decomposition gradually takes place. Store in brown, glass-stoppered bottle.

The specimens evaluated using this technique have all been mounted in "Trans Optic" mounting powder. Both Lucite and the various bakelites were found unsatisfactory for mounting specimens: the ordinary metallographic Lucite because it quickly crazes when immersed in these solutions, and the bakelites because they are slightly soluble and will deposit a film on the specimen. The latter is readily apparent when a cotton swab that has been dipped in Nital is rubbed over the bakelite.

ETCHING TECHNIQUES

When ready to etch a specimen, a small beaker full of each etching solution is poured out. An amount sufficient to cover the specimen about 3/8 inch is put in a small, glass-covered beaker. The covered beakers containing the three etches, one percent Nital, saturated picral, and 20 percent sodium metabisulfite, are put in a constant temperature bath. The temperature of these etches is maintained at 80 ± 2 F. For consistent results, the temperature of these solutions *must* be kept constant and within the indicated range.

The following etching procedure is now undertaken:

1. Remove micro-specimen from final polishing operation, wash with liquid green soap, rinse, and dry with ethyl alcohol and hot air. If the specimen has stood overnight, the final polishing operation should be repeated before etching.

2. Etch the specimen in the 1 percent Nital by immersion. Etch until the structure just begins to become visible, 3 to 5 seconds, then rinse and dry.

3. Etch the specimen in ethereal picric acid with the zephiran chloride by immersion. Etch until grain boundaries are slightly delineated. Time may have to be adjusted to ensure good grain boundary attack. Gently shake specimen while immersed.

4. Neutralize at end of etching time with 10 percent ammonium hydroxide in alcohol. Rinse with alcohol to remove all traces of nitric acid. Dry with ethyl alcohol and hot air. All picric acid must be neutralized before going on to the next etch.

5. Etch or stain with the 20 percent sodium metabisulfite and aerosol. Etching time depends on the temper given to the specimen; in any case etching with this solution is cumulative and may be repeated until the desired contrast is obtained. Start with about two seconds immersion. Once staining starts, care must be taken not to over etch since the process seems to accelerate once it has started.

6. Rinse with hot water (if necessary due to over-etching with the metabisulfite, try wiping with a cotton swab to improve contrast) and then dry with ethyl alcohol and hot air.

7. Specimen is handled with nickel tongs during the etching sequence.

8. The ethyl alcohol mentioned above is pure and undenatured ethyl alcohol from government stores.

9. The above etching procedure has been used for etching HY-80 steel and seems to work best on a tempered structure.

10. If higher nickel content steels, such as HY-140 are being examined, the best etching procedure is to use just the alcohol picric acid etch followed by the Nital etch. On this steel the metabisulfite just seems to cloud over and obscure the microstructure.

INTERPRETATION

When finally etched, the microstructure will exhibit the following characteristics:

1. Tempered martensite will appear grey due to the staining action of the sodium metabisulfite, and will exhibit typical martensitic substructure.

2. The isothermal products will be a bright metallic color since they are untouched by the stain.

a. Proeutectoid ferrite will usually exhibit smooth, even grain boundaries and a flat, unbroken, blocky area.

b. Bainites will exhibit ragged grain boundaries, with this becoming more pronounced at the lower transformation temperatures. In general, it may also be said that the bainites will not exhibit the large flat blocky areas of the ferrite but will be smaller and peppered with darker areas of martensite or precipitated carbides; this becomes more pronounced as the transformation temperature is lowered.

3. A smeared structure or an oxidized surface will make proper evaluation of the results of this technique impossible.

APPENDIX C

PHOTOMICROGRAPHS

OF

REPRESENTATIVE MICROSTRUCTURES

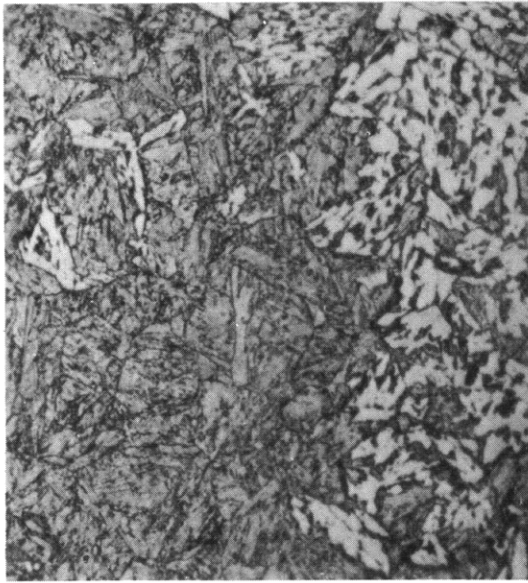


Figure C1a – 29 Percent Bainite – 71 Percent Martensite
875 F – 152 Sec

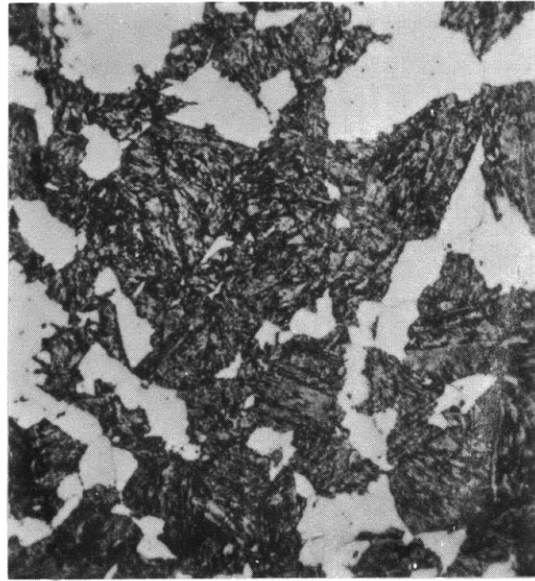


Figure C1b – 31 Percent Ferrite – 69 Percent Martensite
1200 F – 3350 Sec



Figure C1c – 54 Percent Bainite – 46 Percent Martensite
875 F – 1600 Sec

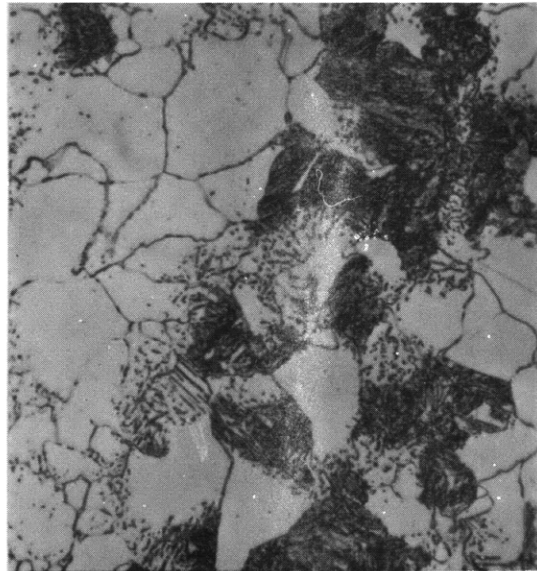


Figure C1d – 67 Percent Ferrite – 11 Percent Pearlite
22 Percent Martensite
1200 F – 8500 Sec

Cropped End Number X-1

Figure C1 – Microstructure (1000X) of Specimens from Cropped End Number X-1 Austenitized at 1640 F, Isothermally Treated at 875 F and 1200 F, and Then Tempered at 1150 F



Figure C2a – 14 Percent Bainite – 86 Percent Martensite
875 F – 152 Sec

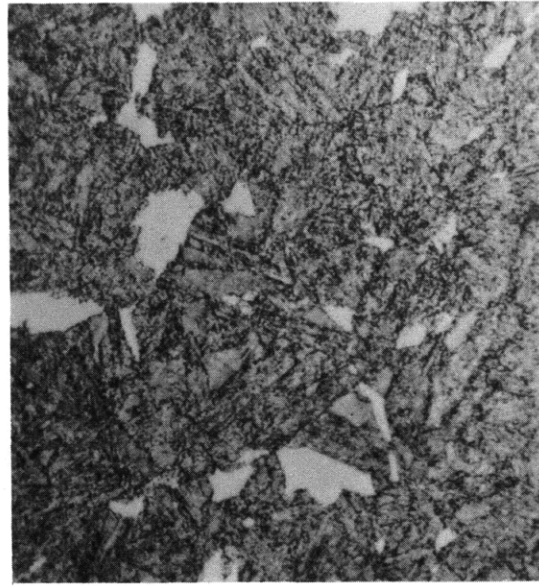


Figure C2b – 5 Percent Ferrite – 95 Percent Martensite
1200 F – 3350 Sec

Cropped End Number X-8

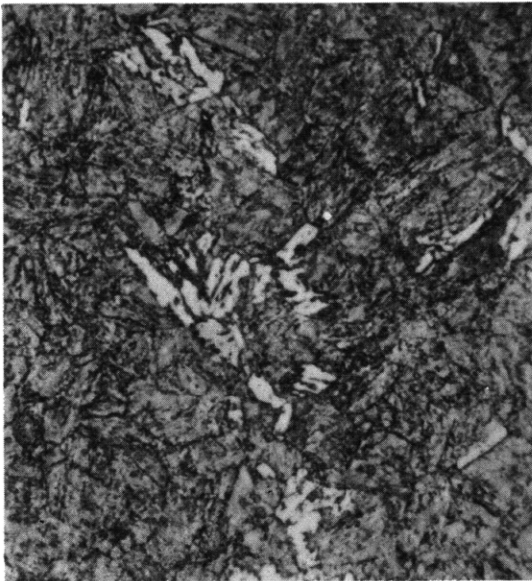


Figure C2c – 7 Percent Bainite – 93 Percent Martensite
875 F – 152 Sec

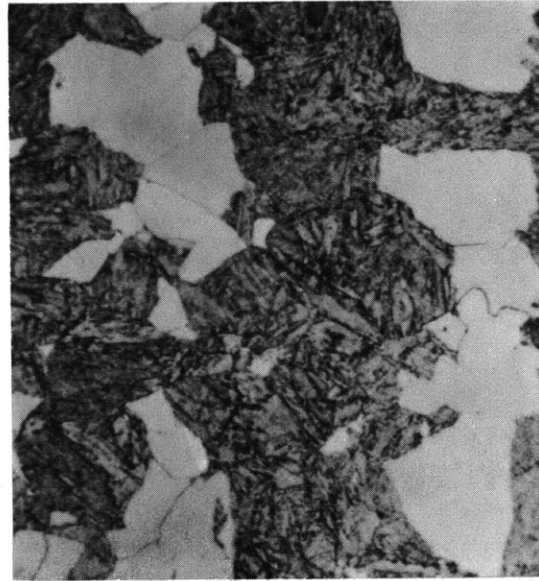


Figure C2d – 26 Percent Ferrite – 74 Percent Martensite
1200 F – 3350 Sec

Cropped End Number X-9

Figure C2 – Microstructure (1000X) of Specimens from Cropped Ends Numbers X-8 and X-9
Austenitized at 1640 F, Isothermally Treated at 875 F and 1200 F, and Then
Tempered at 1150 F

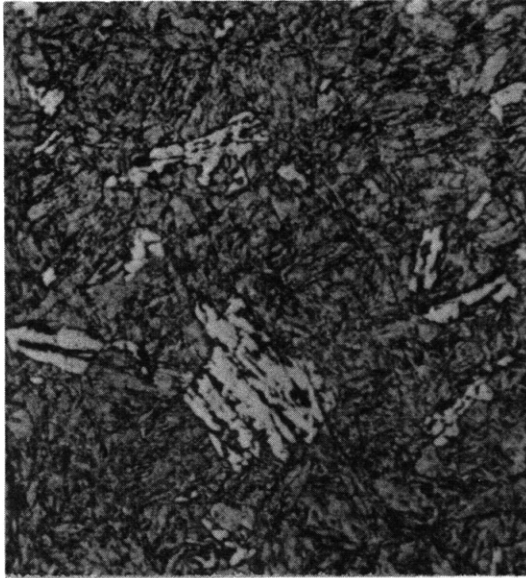


Figure C3a – 9 Percent Bainite – 91 Percent Martensite
875 F – 152 Sec

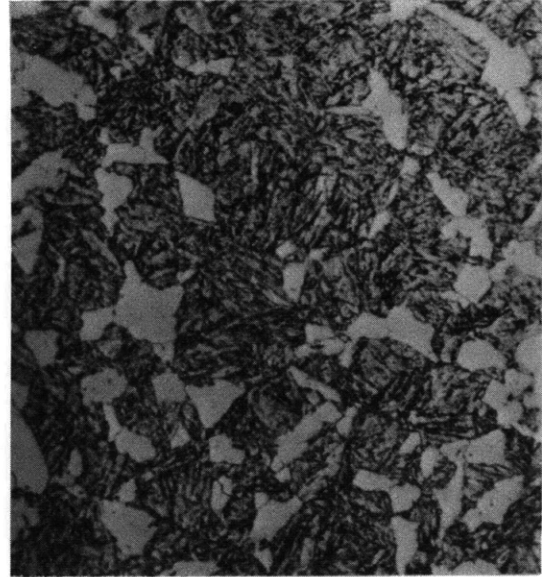


Figure C3b – 22 Percent Ferrite – 78 Percent Martensite
1200 F – 3350 Sec



Figure C3c – 41 Percent Bainite – 59 Percent Martensite
875 F – 1600 Sec

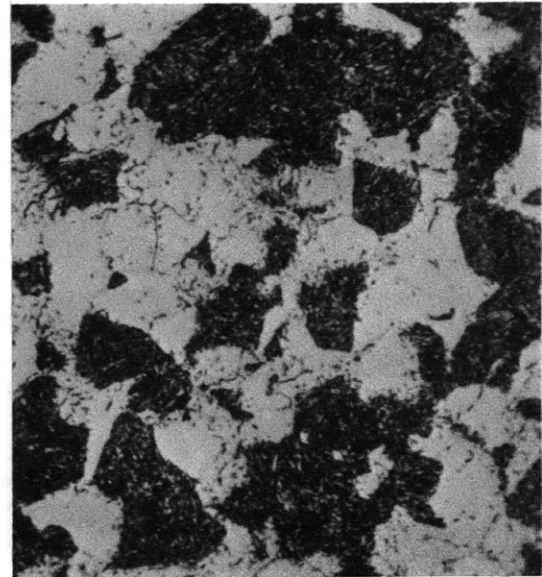


Figure C3d – 53 Percent Ferrite – 4 Percent Pearlite
43 Percent Martensite
1200 F – 8500 Sec

Cropped End Number Y-2

Figure C3 – Microstructure (1000X) of Specimens from Cropped End Number Y-2 Austenitized at 1640 F, Isothermally Treated at 875 F and 1200 F, and Then Tempered at 1150 F



Figure C4a – 50 Percent Bainite – 50 Percent Martensite
875 F – 1600 Sec

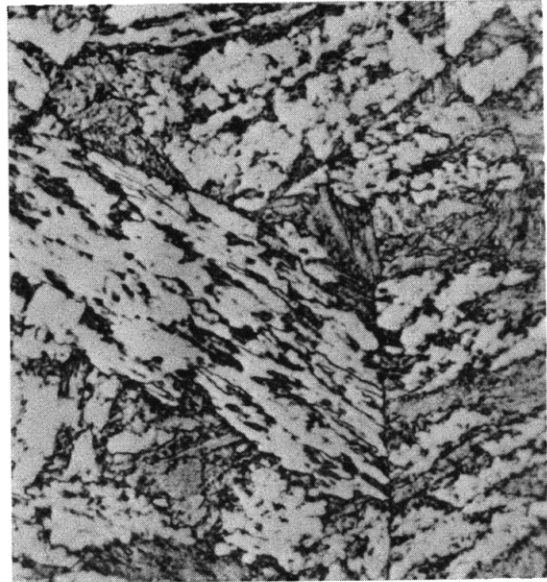


Figure C4b – 63 Percent Bainite – 37 Percent Martensite
875 F – 1600 Sec

Cropped End Number X-12

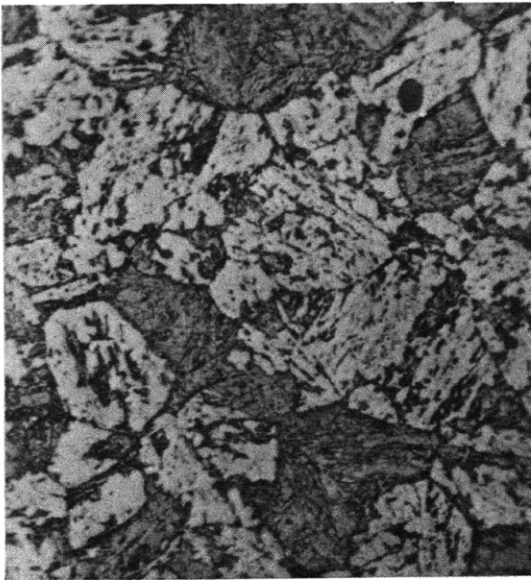


Figure C4c – 60 Percent Bainite – 40 Percent Martensite
875 F – 1600 Sec

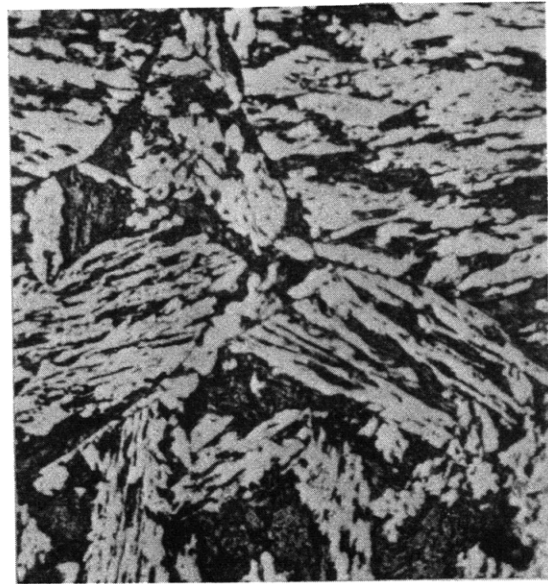


Figure C4d – 67 Percent Bainite – 33 Percent Martensite
875 F – 1600 Sec

Cropped End Number X-13

Figure C4 – Effects of Prior Austenitic Grain Size on the Microstructure (1000X) of
Cropped Ends Numbers X-12 and X-13

BLANK

APPENDIX D

TIME-TEMPERATURE COOLING CURVES FOR DROP WEIGHT SPECIMENS WHEN TRANSFERRED FROM 1640 F FURNACE TO MOLTEN SALT BATHS AT EITHER 1200 F OR 875 F

INTRODUCTION

The actual time-temperature history was desired for the low-carbon Ni-Cr-Mo steel (5/8 × 2 × 5 inches) drop weight impact specimens used in these tests when they were transferred from a high temperature salt bath at 1640 F to a salt bath at either 1200 F or 875 F. In order to determine this thermal history a specimen was instrumented with a thermocouple at its center as recommended by Sinnott and Shyne⁸ and the time-temperature history of the specimen and the bath were recorded at least every 5 seconds on a strip chart recorder. The experimental cooling curve is compared to the curve calculated using the method of Sinnott and Shyne.

SPECIMEN PREPARATION

A 0.087-inch diameter hole was drilled from one end, 2 1/2 inches deep into the center of the specimen. The hole was countersunk and tapped for 1/4–28 threads to a depth of 3/8 inch. A 7 inch length of 1/4-inch high pressure steel tubing was screwed tightly into the hole. An insulated 28-gauge Chromel-Alumel thermocouple was inserted in the tube and forced against the specimen at the bottom of the well. Another thermocouple was positioned in the low temperature salt bath adjacent to the specimen. Both thermocouples were attached to a multichannel recorder to alternately record the temperatures indicated by both thermocouples. The specimen was sand blasted between each run to remove all scale and salt and return it to a representative condition.

HEAT TREATMENT

The specimen was suspended vertically in a high temperature salt bath and allowed to come to 1640 F. It was held at 1640, ± 15 F for 30 minutes, minimum. The ± 15 F bath temperature tolerance was used here and for the low temperature baths because it represented the tolerance of the commercial temperature controller used in this test as opposed to the closer ± 5 F manual control used for the actual specimen heat treatments. From 4 to 5 inches of the thermocouple tubing was kept below the surface of the molten salt to minimize heat losses from the tube.

The specimen was transferred from the 1640 F bath to the low temperature bath within 1 or 2 seconds. The specimen was quenched vertically into the center of the low temperature bath and mildly agitated near the thermocouple recording the bath temperature. Multiple runs were made to allow for slight variations in the bath temperatures between runs.

RESULTS

The results of the cooling rate studies are plotted in Figure D1 and are compared to the salt-bath cooling rate predicted using the instantaneous film coefficient method of Sinnott and Shyne,⁸ with the following parameters:

Specimen Area	0.200 sq. ft. ²
Specimen Weight	1.771 lb
Mean Specific Heat, C _p *	
1640 to 1200 F	0.127 BTU/lb/F
1640 to 875 F	0.128 BTU/lb/F

Table D1 is a listing of the computer program used to predict cooling rates.

The variable names used in this program and their significance are given below:

ITB	Temperature of low temperature bath, 875, 900 or 1200 F
ITS	Initial temperature of specimen, 1640 F
HF	Instantaneous film coefficient based on the difference between the average surface temperature of the specimen and the bath during the time interval THETA
QTH	Average cooling rate in BTU/sec when cooling from MTS1 to MTS2
THETA	Time to cool from a temperature MTS1 to MTS2, seconds
SUMTH	Total elapsed time to cool from ITS to MTS2, seconds

The data shown in Figure D1 indicates reasonable agreement between experimental and predicted cooling rates. The low bath temperature experiments were run with an initial bath temperature of about 900 F rather than 875 F because of the characteristics of the commercial salt mix that was available in the salt bath when these tests were run.

The data shows that it takes about 125 seconds for the center of the specimen to cool from 1640 F to 1210 F and about 140 seconds to cool from 1640 F to 885 F when a drop weight specimen is transferred to a 1200 F and an 875 F bath respectively.

*Based on data from U. S. Steel Corporation.

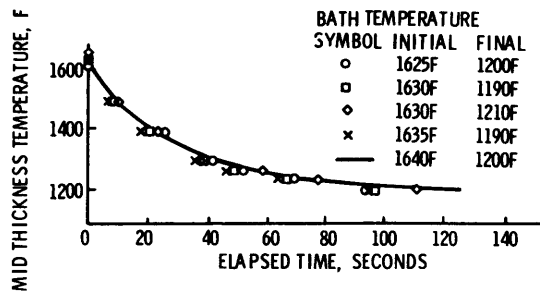


Figure D1a – Cooling Rate from 1640 to 1200 F
(Average $C_p = 0.127$ BTU/lb/F)

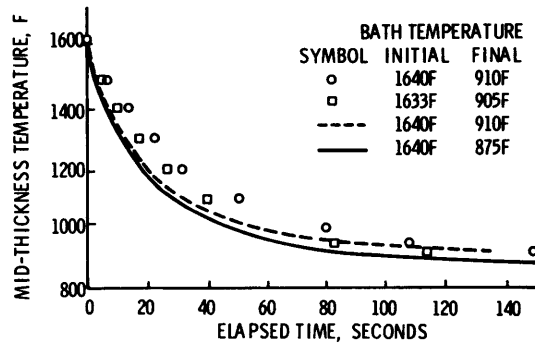


Figure D1b – Cooling Rate from 1640 to 875 F
(Average $C_p = 0.128$ BTU/lb/F)

Figure D1 – Comparison of Actual Cooling Rates with Computed
Cooling Rates for the Center of a 5/8 x 2 x 5 Inch
Drop Weight Specimen

TABLE D1

Listing of the Computer Program Used to Predict Cooling Rates

BUB

```

100 CP=.128
110 ITB=875.0
120 SUMTH=0.0
130 A=.19965
140 W=1.7708
150 ITS=1640.0
160 PRINT,"          ***** AGITATED BY HAND *****"
170 PRINT,"^
180 PRINT," MTS2          HF          QTH          THETA          SUMTH"
190 DØ 20 M=1,80
200 MTS2=ITS-M*10
210 IF(ITB-MTS2)15,20,20
220 15 MTS1=ITS-(M-1)*10
230 HF=93.0*EXP(0.0015*((MTS1+MTS2)*0.5)-ITB))
240 QTH=HF*A*(MTS1-ITB)/3600.0
250 THETA=(W*CP*(MTS1-MTS2)/QTH)
260 SUMTH=SUMTH+THETA
270 PRINT,MTS2,HF,QTH,THETA,SUMTH
280 20 CONTINUE
290 END

```

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c.			
d.			
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13. ABSTRACT <p>Steel from 22 heats of low-carbon Ni-Cr-Mo steel (MIL-S-16216G and ASTM A543-65) were heat treated to study the effects on the strength-differential effect and the difference between tensile and compressive yield strength of (1) commercial variation in composition and inclusion content, (2) variation in microstructure such as prior austenitic grain size and the relative amount of isothermally produced ferrite or bainite in a tempered martensitic matrix, and (3) the observed variation in strength obtained after a 1-hour 1150 F temper followed by a water quench to prevent embrittlement while cooling from the tempering temperature.</p> <p>The difference between the tensile and compressive yield strength, sometimes called the strength differential (S-D) effect, was observed in this study to be at least 5 percent of the tensile yield strength. Data are cited to show that in the low-carbon Ni-Cr-Mo steels studied here, the S-D effect observed was a relatively constant percent of the tensile yield strength, and was markedly structure-sensitive to prior austenitic grain size, microstructural constituents, tempering temperature, type and distribution of carbides formed during tempering, and tempering slightly above the lower critical (A_{C1}) temperature.</p>			

14. KEY WORDS	LINK A		LINK B		LINK C	
	ROLE	WT	ROLE	WT	ROLE	WT
Strength-Differential Effect Tensile Yield Strength Compressive Yield Strength HY-80 Steel Ni-Cr-Mo Steel Microstructure Composition Strength						

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