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STATIC AND BALLISTIC BEND TESTS ON
AN EXPERIMENTAL 2.5" ARMOR PLATE

By A. E. Ruark and H. L. Smith
Ballistics Section - NRL

P. E. Shearin, R. M. Trimble
and Norwood Michie
University of North Carolina

- Report O-2687 -

UNC No. 30

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ABSTRACT

A 2.5" plate of Ni Cr steel, aluminum killed, was supplied by the Carnegie-Illinois Steel Company in blocks which had undergone nine different heat treatments. These were subjected to low speed bend tests for size effect determination. Bend tests at ballistic speed were also carried out and one treatment was subjected to both high and low speed bend tests, in such a way that precracking work and post-cracking work could be determined. The results are these:

(1) Size effect is quite insensitive to the treatment employed, showing little variation with grain size or tensile strength. However, it is markedly increased by temper embrittlement.

(2) The static and ballistic bend results are substantially parallel except in temper-embrittled material and in the products of isothermal transformation at 575 and 625°C. In these three cases the ballistic-bend work-values fall much below expectations.

(3) Experiments on postcracking work were performed on a fine grained tempered martensite. For this material, the ratio of postcracking work to total rupture work is about the same at both high and low speeds. All fractures were ductile in this test series. It is believed that the result is not general, - that the ratio would drop off with increasing speed if a higher hardness level were employed or if tests were made on more brittle material.

(4) In this steel the work density measured in quasistatic tests is not sensitive to grain size, as established by varying the austenitizing temperature. It increases with tensile strength over the range of strengths employed, which was not wide. A 575-degree isothermal transformation product gave work densities almost as high as those for tempered martensite at the same strength level; but, as mentioned before, this good behavior could not be duplicated at high speeds.

(5) Directional effects in the work density in tempered martensite are reduced by a complete re-heat treatment, the mechanism being an improvement in the work density for transverse specimens. Such equalization was not found in a coarse pearlitic structure, formed isothermally.

(6) Simple equations for the rupture-work, and for the post-cracking work per unit area of fracture, have been found, and these are applied to the data obtained on this steel.

(7) The causes of size effect are discussed.

(8) In Appendix I, the chief results of earlier NRL scale effect studies on steels are reproduced, for comparative purposes, and for completeness.

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INTRODUCTION

Authorization

1. This work was authorized under BuOrd Project No. 40025 and NRL contracts N173s-8171 and N173s-9974.

Statement of Problems

2. The general recognition that poor quality in armor plate is often connected with brittle behavior at the back of the plate has led to investigations of many causative factors. The effect of hardness was well explored in early work at NRL (Ref. 12). Now it is well recognized that increase of hardness, size, strain rate, and severity of stress-pattern, as well as decrease of temperature, are all factors in the production of brittle behavior. This report is concerned primarily with the effects of size, strain rate, and direction of the maximum principal stress. All these effects are ultimately dependent on microstructure and hardness level. It was expedient to avoid a host of uncertainties by using a number of different treatments of a single steel.

3. In earlier reports (Refs. 1 and 2) we have described accurate equipment and techniques for studying the size effect by means of quasi-static bend tests on geometrically similar specimens of square cross section. In these tests the load-deflection curve is determined autographically. The area under this curve measures the work absorbed by the specimen. In this investigation, attention will be focused on the total work required to rupture the specimen, the work absorbed up to the instant of maximum load, and the work absorbed after the maximum load has been passed.

4. It is convenient to divide each of these works by a number proportional to the volume of the specimen, thereby obtaining a quantity which is called the work density, W . The advantage is that if there were no size effect, the work density would be the same for specimens of all sizes. Therefore the variations of W furnish a measure of the size effect.

5. This report presents the results of bend tests, at both quasi-static and ballistic speeds, on a single carefully manufactured experimental armor steel, available in a number of heat treatments. The specific points to be examined are the changes in the rupture work associated with changes in size, testing speed, and heat treatment. The following factors, varied by means of heat treatment, have been studied.

(a) Influence of tensile strength in a series of fine-grained tempered martensites.

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- (b) Influence of grain size, at constant tensile strength.
- (c) Influence of temper brittleness, at constant tensile strength.
- (d) Comparison of fully quenched and tempered steel with steel transformed isothermally; all specimens being tempered to the same hardness, and all being coarse-grained.
- (e) Effect of variations in the temperature of isothermal transformation, for coarse-grained products.

Earlier Work

6. Size effect results for a number of armor steels and for 24ST aluminum have been given in numerous progress reports from the University of North Carolina. Work has also been done on size effect in the tensile test as an indicator of size effects. The entire body of work up to 1 July 1944 has been summarized in two reports, Refs. 1 and 2. Bend tests of similar character performed at the Carnegie Institute of Technology are also summarized in Ref. 2.

7. The Carnegie investigators used a deck armor steel and established the general nature of the variation of size effect with hardness. For their steel, the work density plots as a straight line of negative slope, against hardness, on log-log paper. Also they find that the logarithm of the work density is a decreasing linear function of log (size), for constant hardness. The slope of the work-versus-size curve on log-log paper, taken without regard to sign, is called the size effect index. In our work, covering a variety of armor steels, this linear relation has seldom occurred. We have found that the experimental situation can be well represented by two simple empirical equations, presented in Section entitled "Models and Equations Describing the Size Effect", page 23.

8. Our early work on this project was exploratory, because of the almost complete lack of size-effect data. It was necessary to study a number of factors which influence the size effect. Tests were made on material from plates of greatly different thickness. Steels which varied considerably in regard to cleanliness were studied. An attempt was made to assess the importance of temper brittleness. By using 18" armor, it was possible to work separately with tempered martensite from locations near the face, and with tempered bainite of surprising homogeneity, from the plate interior. Also it was possible to study directional effects, using specimens whose axes lay in all of the three principal directions in the tempered bainite. Selected results from this earlier work will be compared with the conclusions from the present investigation in Section entitled "Discussion of the Data", page 16. The summary-section of NRL Report O-2412 is reproduced in Appendix I to give a complete review of earlier work.

Original Work Done

9. Late in 1943 it became clear that systematic tests of a single armor steel in various heat treatments would be the most logical procedure. On 12 January 1944 representatives of NRL and of UNC conferred in Pittsburgh with Dr. E. C. Bain, Mr. John Hodges and Dr. W. H. Mayo of the Carnegie-Illinois Steel Corporation. Arrangements were made for the purchase of a 2.5" plate from a special heat of Ni-Cr steel, killed with aluminum. Detailed plans were made for re-heat treatment of a number of subplates, so that various microstructures and physical properties would be obtained. Actually, 11 blocks in 9 different treatments were deemed sufficient for all the purposes in view. The primary purpose was to secure material for size effect studies by means of the bend test, but provision was made for an ample supply of as-received material, for other tests. (In this report, the words, "as-received material", refer to the portions of the plate which have not undergone any special treatment after the original quench and temper.) A suitable plate, No. 52151, was manufactured and treated under the supervision of scientific and technical personnel of the Carnegie-Illinois Steel Corporation. Using this plate, the size effect has been studied in quasi-static bend tests, hereafter called static tests. All the 9 treatments have been subjected to bend tests at ballistic speeds, using a single specimen size. A special series of experiments was made to determine the precracking and postcracking work, for as-received material, at both slow and high speeds.

EXPERIMENTAL METHODS

Static Bend Tests

10. For static bend tests the specimen is fastened in two massive holders (Figure 1) having suitable square sockets to receive it. The load is applied to the assembly at two points by means of ball or roller bearings, and the assembly is similarly supported.

11. Figures 2 and 3 show a bending apparatus, set up on the weighing table of a Richle 100,000 pound testing machine. The rugged construction of all parts is apparent. In Figure 3 the specimen holders are inclined, because the photograph was taken in the late stages of a test on a specimen 1.875" square. The developing crack in the specimen can be seen in the space between the holders. Above, one sees a carriage carrying the upper rollers. The small unit at the top having several electrical connections, is a compression column of high-strength steel carrying a set of resistance strain gauges, (Types A-1 and A-7). The compression head of the testing machine rests directly on this column, and the elastic strain of the column gives a direct and accurate measurement of the applied force. A second column used for compensation of thermal effects in the strain gauge circuit is also shown.

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12. Figure 2 shows benders for use with four specimen sizes. Equipment is available for 5 sizes, viz. $3/8$, $6/8$, $9/8$, $12/8$, and $15/8$ inch, referred to as sizes 1, 2, 3, 4, and 5. Only the sizes 1, 3 and 5 were used in the present work.

13. A simple plan was developed for automatic and accurate recording of curves giving reduced load (load/retained area of specimen) versus reduced deflection (deflection of loader/retained depth of specimen), in such a way that if there were no size effect the curves obtained for all specimen sizes would be exactly alike. The principles involved are as follows:

(a) The testing machine head is driven at speeds proportional to the specimen size, so that in the absence of size effect plastic strain rates would be held constant. This is accomplished by an induction motor and a Link-Belt positive infinitely-variable speed-changer. The load on the motor is small and it operates at practically constant slip speed. The total head deflection in a given time is checked with a traveling microscope. The recorder used has a paper chart driven by a synchronous motor so that paper motion is proportional to the deflection of the specimen. The error due to variations of paper motion is negligible. The head speed is 0.0002 in. per sec. for the No. 1 specimen, 0.0006 in No. 3, and 0.001 for No. 5. Testing times range from about 30 min. for $1-7/8$ " specimens to 100 min. for $3/8$ " specimens.

(b) For recording the load, we employ the output of the resistance-gage bridge to feed a Leeds and Northrup potentiometer recorder, with 10 millivolt scale. Four gages, in two parallel sets, form two active arms of the bridge, and four others attached to a similar steel column are used in the two remaining arms for temperature compensation, as seen in Figure 3. Gages were later mounted on a one-inch steel plate for the No. 5 specimen. Test runs showed a variation from linearity of load versus deflection well under 1%, and successive loadings and unloadings showed a spread of less than 1%.

(c) Column sizes and currents were so chosen, for each specimen size, that the recorded loads were in scale; that is, if there were no size effect the curves for all specimen sizes would be identical. The reader may turn to Section entitled "Ballistic Bend Tests", page 6 for further explanation.

(d) In this work changes of the chart dimensions due to humidity must be taken into account. This is done for the displacement axis by measuring the total travel of the testing machine head during each test with a traveling microscope (Figure 2), and by carefully indicating the starting and stopping points on the paper. Thus a known head travel corresponds to a known paper distance. Likewise, for the load axis, the

beam of the testing machine is balanced manually at the beginning and end of the test, and at the maximum load. Thus the load indicated by the recorder is calibrated in each test.

14. All specimens are geometrically similar.

Sizes and Tolerances for Square Bend Specimens (inches)

Nominal Size	3/8	1-1/8	1-7/8
Notch Diameter,	0.023 $\pm$.001	0.068 $\pm$.002	0.114 $\pm$.002
Width of Specimen,	0.375 $\pm$.005	1.125 $\pm$.100	1.875 $\pm$.010
Gross Depth,	0.375 $\pm$.000 - .001	1.125 $\pm$.000 - .002	1.875 $\pm$.000 - .002
Retained Depth After Lapping,	80% of Gross Depth $\pm$ 2%		

15. The specimens are brought to size by shaping and grinding. Special care is bestowed on notch preparation. The 90 degree notch is terminated by a semi-cylindrical portion. It is cut in a milling machine, and the semi-cylindrical portion is lapped to size with drill rod and abrasive, eliminating the milling marks in the notch direction; these would be effective small notches themselves, which would influence the behavior of the specimen under test.

16. The coordinates we use in describing specimens are these; OX is the rolling direction, or, in cast material, the ingot direction. OY is the direction at right angles to the rolling direction, in the plate face. OZ is perpendicular to the plate face. The notch is cut perpendicular to the plate face.

Ballistic Bend Tests

17. Figure 4 shows a 50-caliber smooth-bore machine gun barrel, arranged to fire through a silencer and blast suppressor into a ballistic pendulum. Figure 5 shows the pendulum and recording attachments, with pendulum cover raised so that the reader can observe a specimen mounted in a steel jig, ready to be ruptured. Figure 6 shows the special projectiles employed. They are hardened steel cylinders with spherical ends, weighing 1000 grains or 1/7 pound. Specimens, both ruptured and unruptured, are also presented. The specimens are 1/2"

thick, 1-1/2" wide and 4" long. They are notched on all 4 sides, using an Izod cutter to produce a notch 3/32" deep so the retained area, amounting to 0.410 sq. in., has dimensions 5/16 x 21/16".

18. The ballistic pendulum measures the momentum of the projectile because all parts inside its case come to rest in a time short compared with the period of swing. Knowing the bullet mass, the initial bullet energy can be computed. In the ideal case, the specimen would be barely severed and the residual kinetic energies of the specimen and the bullet would be zero. Three to six shots usually suffice, in the hands of a skilled observer, to give a limit velocity satisfactory for present purposes. Variations in the material set a limit to the amount of effort which should be expended in locating a limit velocity. It is relatively easy to find a narrow velocity range in which the residual energy of the specimen-top is negligible, but not so easy to make the residual bullet energy zero at the same time. In practice, the residual bullet energy is made quite small by properly choosing its mass, relative to that of the moving part of the specimen. In presenting results, no correction is made for the residual energy. We do not believe that the general picture of the ballistic bending behavior of these steels would be appreciably altered if such corrections were introduced. Experimental errors are discussed in Section entitled "Data", paragraph 44 page 14.

19. The bullet energy is utilized in several ways. There is a part E_b which produces plastic flow at the point of impact. Some dissipation of this kind is unavoidable in tests at such high speeds. It has been brought to as low a level as is conveniently possible, by proper design of the projectile and the specimen. Another part, E_w , escapes to the mounting in the form of elastic wave energy. Another part E_n serves to separate the specimen at its notched section. Thus the test is not a direct measure of the rupture energy, as was the case for the slow bend test. To a very rough approximation, the fraction of the kinetic energy used in the neighborhood of the notched section is

$$E_n/0.5 mv^2 = 1.33 m/M, \quad (1)$$

where m is the bullet mass and M the specimen mass. This expression is derived in Ref. 4. The derivation assumes that the specimen behaves like a rigid uniform bar, struck at its center of percussion, and that the damage at the point of impact is completely localized, so that the bullet energy is handed over to the specimen before it has time to move appreciably. These assumptions are inaccurate, but equation (1) is useful for purposes of design. A mathematical study of the real situation reveals that an accurate determination of E_n would require a knowledge of the torque applied to the portion of the specimen below the notch during the breaking process. This torque would have to be measured by the use of dynamic stress gages, applied to the specimen at a point very close of the deformed region at the notch. It is not practical to do this at present, at the high speeds employed in these experiments, so we have preferred to forego any statements as to the value of E_n , and to focus attention on the determination of the bullet

energy required for rupture. Equation (1) indicates that the fraction of the energy utilized at the notch can be raised by increasing m/M . In practice, m/M is limited to about 0.44, in order to avoid follow-through of the bullet, leading to high values of the residual kinetic energy. Even with this m/M value the projectile continues to travel forward, after its first rebound from the specimen, which itself is moving forward at the instant of rebound. The rotational energy of the specimen decreases as it bends, since work is being done at the notch, so the bullet catches up with the specimen again, striking it an additional weak blow, or several blows. This is really an advantage, for the small imprint made on the specimen during the second impact furnishes a good means of making sure that most of the energy has been lost in the first impact. Furthermore, the additional blows use up additional bullet energy, reducing the residual energy to proportions which can be neglected.

20. To locate the domain of strain rates employed in these tests, we need a rough formula for an average strain rate at the notch at the beginning of the bending process. To get an estimate, let us treat the specimen as though it were a uniform bar; but let us assume that it is made of two materials, so constituted that flow is confined to a short length at the middle, - a length $2X$, equal to the root diameter of the notch, or 0.5mm. Then it turns out (Ref. 4) that the initial strain rate in this short length is

$$de/dt = \frac{VD}{4L_p X}, \quad (2)$$

where the symbols have the following meaning:

V = initial velocity of the point of impact.

L_p = distance from notch to point of impact (1.37" or 0.115' in our case.)

D = depth of the beam, assumed equal to the retained depth below the notch. (5/16" in our case.)

In these experiments, with V in feet per second and lengths in feet, we get

$$de/dt = 34.5V \text{ inches per inch per second}$$

The values for several velocities are roughly as follows:

V in fps:	500	1,000	1,500
de/dt in sec ⁻¹ :	17,000	34,000	51,000

The uncertainties of this calculation are such that these values should be used only for rough comparisons. As shown in Table 5, the speeds

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employed in this work are of the order of 1000 fps, the highest being 1400. As far as we can find these are the highest rates which have been used in any mechanical test, aside from usual plate-penetration tests. There will be no difficulty about obtaining higher rates in bend tests of still stronger steels. By way of comparison, the highest rates we have found in the literature are about as follows:

In tensile impact tests of notched specimens: 16000/sec.
In compression impact tests with wheel-type machines: 1600/sec.
In tensile impact tests of unnotched specimens, carried out with a wheel-machine at California Institute of Technology: 300/sec.

21. It is clear that this type of ballistic bend test is adapted to give us information under well controlled conditions, using a relatively simple type of stress and strain pattern, in the region of strain rates where our ignorance is greatest; namely, from about 300 inches per inch per second up to the strain rates encountered in conventional plate penetration tests. By using projectiles of different masses it is possible to look into questions of speed sensitivity in the ballistic range of velocities. The range of speeds which can be covered with any given type of specimen is not wide, but by using several types the range can be extended step by step.

Measurements of Precracking and Postcracking Work

22. It is felt that a much better understanding of the rupture of metals can be obtained if the work required to initiate cracks and the work required to extend the cracks to completion are separately measured. It is well known that the work associated with crack propagation is expended in producing plastic flow, the surface tension being only 10^{-3} to 10^{-4} ft. lbs./sq.in. Portions of the specimen at some distance from the advancing crack head are free to behave much as they were behaving when the crack had not yet appeared. The difference between the precracking and postcracking regimes of deformation lies in the fact that the volume involved in flow is much more restricted in the latter case. Another difference is that the average strains are higher after cracking begins.

23. It is clear that when the ductile type of fracture is replaced by brittle fracture, the drop in work may be, and often is, mainly associated with the postcracking phenomena. It is regrettable that values of precracking and postcracking work have not been determined long ago for a variety of steel specimens. The only useful data we have found are due to Docherty (Ref. 5). These have been reviewed in Ref. 4. Time did not permit a complete study of this matter for all nine treatments of steel 52151, but we undertook to determine the precracking work, E_a , and the postcracking work, E_p , for the as-received condition. These quantities are dependent on specimen proportions and size, but of course this is true of all the so-called mechanical properties of a metal. Since both quantities may depend on the speed of straining, both static and dynamic bend tests were made, using the

1.5" x 0.5" x 4" specimen with Izod notch on all 4 sides, as described above. The procedure for the static bend tests is simple. An optical arrangement for viewing the notch at a magnification of 60 was employed, to find the instant at which the cracks in the notch attained a certain total length. The work done up to that time was arbitrarily defined as the precracking work, E_a . At first sight we might think it better to note the instant at which the first visible crack appears, but this is not a good procedure. Experiment shows that surface flaws at the notch give rise to numerous small cracks and that many of these may come to a halt, remaining substantially unaltered until the specimen has experienced considerable increments of strain. Furthermore, it is possible that cracks may form in the solid metal before they become visible at the notch surface, because of triaxial stresses, which are most severe a short distance beneath the root of the notch. Irregularity in the occurrence of the first minute surface cracks makes it advisable to choose a later stage as the termination of the first stage of destruction. The instant at which the cracks extended across 90% of the notch length was chosen as the termination of the first stage. In this steel, the small initial cracks do not join up to form a single main crack at this stage; individual cracks have already penetrated a small but unknown distance into the specimen, this distance being greatest at the notch center. Thus it is somewhat of a misnomer to call the work, E_a , the precracking work, but we shall continue to use this convenient term, since the above explanation precludes misunderstandings.

24. In the ballistic bend tests, we first used a group of specimens to determine the total rupture work. Others were precracked to the "90%-of-notch-point" by means of the static bend apparatus, and were then reduced in length to fit our ballistic testing accessories. This group was then broken to determine the bullet energy necessary to complete the severance.

25. The results of the static and ballistic tests are not directly comparable. As explained in Section B, we could not determine the energies supplied to the region near the notch. We content ourselves with the bullet energies for the two stages of deformation, T_a and T_b . The absolute values obtained in this way are not comparable with the static values E_a and E_b , but it is felt that the ratio T_b/T_a can be profitably compared with the ratio (postcrack work)/(total work) measured in the static tests.

Computation of Work Densities and Size Effects

26. Figure 10 is redrawn from autographic records of slow-bend tests. Let b be the specimen width and d the depth retained at the notched section. Then, in the curve, the ordinate is the load F over the specimen area bd , and the abscissa is the displacement x of the loader divided by the depth d . Therefore the area under the curve is

$$W = \frac{1}{bd^2} \int F dx = \frac{E}{bd^2}, \quad (3)$$

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where E is the rupture-work. W will be called work density since its dimensions are those of work per unit volume. It is understood that it is not the work done in a particular volume but merely the whole work divided by a volume bd^2 . As mentioned before, this quantity would be independent of specimen size if there were no size effect. Actually it is not, and its variations from one size to another constitute a simple measure of the effect of size. In previous reports we have usually employed the ratio of the W-values for two specimen-sizes to show the magnitude of the size effect. For example the ratio

$$W_{0.375}/W_{1.875}$$

is employed to compare the results for the 3/8" and 1-7/8" specimens. At times we have used another plan. Plotting W against a specimen dimension L, on log-log paper, the actual slope of the curve is

$$d\log W/d\log L = (dW/W)/(dL/L).$$

This quantity is a measure of the size effect at a given size level, L. It is practically the percent change of W produced by a one percent change of the specimen dimensions. However, we shall not use it in this paper, since only three specimens were employed and there is little point in measuring the slope of a curve defined only by three points.

27. We now consider the problem of getting the work which has been used to produce plastic flow at any stage of the test, and the problem of finding the work involved in brittle fracture. In Fig. 15, the top graph shows a load displacement curve. When the displacement is OP, the whole work done on the specimen is the area OAP, but if we unload the specimen we recover an elastic work indicated by the area of the triangle ABP. Thus the plastic work done before and after the displacement P is reached are the areas to the left and right of the sloping line AB. The reader will observe that when we carry the specimen on to complete ductile rupture, the elastic energy is transformed into energy of plastic flow; that is why the area ABP represents the plastic energy plowed in after point P is reached. In the top diagram the boundary line AB is drawn parallel to the initial part of the load curve OQ, which corresponds to the modulus line in the diagram for a tensile test. In actuality there is hysteresis in loading and unloading; but this has only a slight effect on the work values. Loading and unloading tests show that the unloading line and the initial rise have slopes which agree within one degree, on the particular scale of plotting used in this work; the effect of a one degree error on the value of work density is of the order of 0.4 ft. lb./cu. in. The data obtained in such unloading tests have given a background of experience for drawing the line AB in cases where it is inadvisable or inconvenient to unload the specimen.

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28. The third diagram in Figure 15 shows what happens when the specimen has a band of brittle fracture across its center. The load falls off suddenly and the energy required to spread the fracture is not reflected by the recorder, which is engaged in its accustomed task of indicating the load and the position of the loader. In short, the energy required to spread the fracture has been supplied by the relaxation of the specimen, the grips and other parts of the bending mechanism. The elastic energy of all these parts was ABP before the fracture and afterward it is A'B'P, so the release of energy is ABB'A'. The work represented by this quadrilateral can be divided by the area of brittle fracture to obtain a value for the brittle fracture work per unit area. When the specimen breaks completely through in brittle fashion this cannot be done. Then we only know that the elastic energy ABP was released, and that this amount is more than enough to rupture the specimen.

29. One last point; specimens of ductile steel cannot be broken completely on our apparatus although it permits a bending angle of more than 90 degrees. For this reason, we have standardized a point at which to terminate the test. This is done when the load falls to 5% of the maximum load. Therefore the statically-determined W-values in this report are really the work-densities required to carry the specimens to this stage of destruction. In comparing similar steels, the omission of the tail of the curve is unimportant.

DATA

Description of Carnegie-Illinois Plate 52151

30. The manufacturing record for this plate is given in Appendix 2, with ladle and plate analyses. The portion received by NRL is from the No. 2 cut of an ingot weighing 32000 lbs. (No. 1 cut is the discard). The composition is substantially that of heavy Class B armor, but aluminum was added to the ladle since fine grain was desired in several of the treatments. Referring to Figure 7, NRL received two sub-plates 36" x 65", labeled S1 and S2. These are called the "as-received" condition. Along each border of the main plate, blocks were cut for further heat treatment by Carnegie-Illinois. Of these, NRL received the five blocks labeled A to E, and six others, subjected to isothermal treatments, which are labeled H to N.

31. The treatments and physical properties are described in Table 1. It was expected that since the thickness is only 2.5" and the alloy content is high, it would be possible to transform this steel completely to tempered martensite, in spite of the fine grain. Inspection of microphotographs indicates that this objective was not entirely attained, but it was substantially accomplished, so we shall speak of certain products as tempered martensite. The physical properties, together with a knowledge of the heat treatments, are deemed more important than a complete description of the microstructures.

32. First, S1 and S2 were annealed, tempered and quenched, to establish a homogeneous fine-grained material suitable for retreatment. Blocks A, B and C were treated to give fine grained tempered martensite at three strength levels. The values sought for and those attained will be quoted:

Desired:	100000	110000	120000
Attained:	101300	110500	117100

33. For comparison with the fine grained products, blocks D and E were treated to coarse-grained tempered martensite, at a tensile strength of about 110000 PSI; block E was then made very temper brittle.

34. The remaining blocks were austenitized at 1100°C and were transformed in pairs at 400, 575 and 625°C. The intention was to make longitudinal specimens from one block of each pair and transverse ones from the other. Unfortunately, blocks J and K were miscut, so transverse specimens of the 400 and 575° products could not be made.

35. A hardness survey through the thickness (14 positions) gave the following results for a sample of treatment s:

Rockwell G hardness	81.3
Average deviation of single result from the mean	0.34
Total spread	80.2 - 82.2

36. Table 7 gives Brinell numbers for individual specimens from each of several treatments. Hardness data for all treatments are summarized in Table 3. The figures for the as-received material in Table 7 serve to show the general magnitude of variations over the plate area. They range from 229 to 237 but within any subplate the variations are sufficiently small to give us confidence that the plate is well suited for our purpose.

37. Figure 16 is an ammonium persulphate macroetch of the material. As compared with typical armor compositions etched under the same conditions, it indicates a fairly clean steel. In view of the low sulphur and phosphorus values it was thought unnecessary to make prints showing the location of these elements.

38. Figure 8 shows typical fractures of several treatments. In the as-received materials, S, almost every specimen has subsidiary cracks perpendicular to the main fracture surface. These give rise to the characteristic lips seen in Figure 8, which are similar to the shear surfaces at the side boundaries of the fracture. The re-heat treatments do not show these cracks which, from their size, are assumed to be caused by planes of sulphide inclusions. We have no other line of evidence at present to support this interpretation. We have never found these lips on cleavage fractures, a statement which applies not only to this steel, but to all others so far tested by the same technique. In Figure 8 the fractures S, B, H, and K are completely ductile. In

the 625° microdict (M, N) we see that the fracture is half ductile and half brittle in the OX specimen and about 90% brittle in the OY specimens.

39. Figure 9 portrays the structure of the fractures found in specimens of treatment E, which is temper brittle. Some OX specimens have a small percentage of ductile fracture surface. This is indistinguishable from the ductile fracture of treatment D, which was not made temper brittle. The remaining area is interesting, because it is almost entirely an intergranular fracture. These photographs were taken with a 5.6X objective, and flat areas had to be selected to get even half of the field in approximate focus. The arrow in the upper-right-hand frame shows the outlines of two dark regions representing prior austenite grains of typical size. The grain size appears to be between 6 and 7. The upper-left-hand frame shows some large well-formed austenite grains, especially in the lower left quarter. The ductile fracture does not reveal austenite grain size clearly because of deformation within the individual grains. Temper embrittlement has so weakened these specimens that subsidiary cracks perpendicular to the main fracture surface are a regular occurrence.

40. Figures 17A and 17B present photomicrographs of five treatments: the as-received material S, the fine-grained tempered martensite B, and all three isothermal products. These photographs show sections normal to the plate surface and parallel to the rolling direction, at or near the center plane of the plate. The photographs at 100 X all include a few sulphide stringers, to show their general nature and to indicate the rolling direction. Banding is clearly seen in the as-received material and visual observation shows that it is present in all retreatments although it is not apparent in all the prints. The photographs at 1000 X indicate that S and B are substantially tempered martensite, although visual observations indicate small amounts of bainite in these center specimens. The treatment of the 400° product H was designed to yield 50 per cent transformation before the final quench to room temperature. Accordingly we see in the microphotograph a white and a dark constituent. Ferrite is clearly visible at both magnifications in the 575° and 625° fractures; the fields photographed at 1000 X were chosen to show slantwise pearlite to good advantage.

Results of Static Bend Tests

41. Results for all specimens are given in Table 2, so that variations of behavior can be studied in detail by anyone interested. In Table 3 the tests are summarized. The quantities recorded are:

W_1 , the work density up to maximum load;

W_2 , the work density after maximum load;

W , the total work density.

The first two quantities have been corrected for elastically-stored energy in the fashion explained in Sec. II D; thus they represent energy densities expended in producing plastic flow. The column headed

"% deviation in W" refers to the expected average deviation of the mean value of W, and not to the average deviation of a single specimen from the mean. The next three columns contain the size effect ratios, the final results desired. It has not been thought worth while to tabulate size effect values for W_1 and W_2 , however, we have provided Figures 11, 12 and 13, showing plots of W_1 , W_2 and W, respectively. In each case the abscissa is the edge of the specimen cross-section.

Results of Ballistic Bend Tests

42. Table 4 reports the results of ballistic bend tests carried out on a single specimen-size. These results may be compared with the results for any size employed in the static tests, but we have arbitrarily chosen to compare them with the 3/8" static specimens. This is done in Table 5. Since the results are extensive they are plotted in Figure 14. In this graph, the scale for bullet energies and that for static work densities have been chosen so that the ordinates of the ballistic and static curves are almost the same for the block A, which yielded exceptionally high values in both types of test. With this arrangement of the scales, speed-sensitivity shows up clearly. If a material is speed sensitive the ordinate of the ballistic-test curve falls below that for the static-test curve.

Results on Precracking and Postcracking Work

43. These results are given in Table 6, which includes both static and ballistic tests. We were agreeably surprised to find that the static values of precrack and postcrack work, E_a and E_b , are fairly uniform. This can be seen from the plus-and-minus values appended to the means in the table. In the static tests the average deviation of a single result from the mean is 6.5% for E_a and only 3% for E_b . Maximum loads are also quite uniform. It is clear that the specimen with notches on all four sides has greatly suppressed the work input before the crack extends across 90% of the notch. The test is primarily a test of postcrack work.

Errors; Spread Due to Inhomogeneity of the Material

44. First we consider errors and spreads in the static tests, where instrumental accuracy is high. Consideration of the errors in calibration and in measurement leads to the belief that the overall experimental error in an individual work value is ordinarily not larger than 0.5%. The intrinsic variations from specimen to specimen are larger. The details can be seen from the column of deviations in Table 3. The actual numbers given are deviations of the means, in percent, as calculated from the formula,

$$100 (\text{Average deviation for a single specimen}) / (n-1)^{1/2},$$

where n is the number of specimens in a group.

45. The W-value for a single specimen shows an average deviation from the mean varying from 0.5% to 20.7%. What is the meaning of this spread of deviation-values? Very small deviations are due to lucky chance. Very large ones occur when some of the specimens in a group show ductile fracture while others show cleavage fracture; they also occur when mixed fractures are encountered. Illustrations are found in the results for treatments D and E, the latter being temper-brittle; also in the treatments M and N; M shows mixed fractures, while N shows cleavage fracture. Omitting these exceptional cases, the deviation for a single specimen has the following average values:

For OX specimens:	3.0%
For OY specimens:	2.8%

46. These figures justify our statement that instrumental errors can be neglected in comparison with specimen variations. As one would expect, they are in line with variations encountered in our precise tensile tests of armor steel.

47. If we increase the number of specimens subjected to a test, the spread of values should increase. The average deviation of a single sample from the mean should approach a constant value, and the estimated error of the mean itself should decrease. It was thought worth while to make a long series of tests on a single treatment of this steel for comparison with the results on the customary number of samples. Treatment C was chosen and 12 additional specimens of 3/8" size were broken, giving a total of 16. The results in Table 2 yield the following survey of the situation:

Quantity Measured	4 specimens (C7X-011X)	16 specimens (C7X-C23X)
Mean work density, ft.lb./cu.in.	2008	2037
Spread of work density	1877-2096	1877-2104
Average deviation of a specimen	4.2%	2.2%
Average deviation of the mean	2.1%	0.6%
Average deviation of the final bend angle	2.8%	3.2%

48. In going from 4 specimens to 16, the mean shifted by 1.4%, which is well within the error limit, 2.1%, assigned to the mean of the first 4 specimens.

49. The data on cleavage fractures and on mixed types are not sufficiently extensive to justify a study of mean deviations. However, several long series of impact tests, (Ref. 7), indicate that scatter is greater when the fracture is by cleavage than when it is of the ductile type. To summarize: the deviation of a single result from the mean is of the order of magnitude 3% in the static tests when the fracture is ductile.

50. There is no guarantee that the same figure holds for the ballistic tests, using a different specimen type. Our velocity determinations are accurate to 0.4%, this error being about equally divided between the readings for a single determination and the error involved in calibrating the ballistic pendulum. This estimate is liberal, (Ref. 6), so experimental error in the individual energy measurement need not be considered in comparison with the errors caused by material-variations, and by the difficulty of seeking the limit-velocity. In determining a limit velocity, the procedure is to continue shooting until there is only a narrow gap between a higher velocity which causes rupture, and a lower one at which rupture does not occur. If sufficient specimens are attacked, crossovers usually occur. These are cases in which one specimen ruptures at a relatively low speed while another fails to rupture at a higher speed. It is our custom to place the limit velocity at the center of the "bracket", or of the crossover range, whichever is higher. The confidence limits assigned in Table 4 are the half-width of the bracket, or of the crossover range, whichever is higher. They yield the following average confidence limits for the energy, not the speed:

Specimens with ductile fractures:	2.4%
Specimens with cleavage, mixed or intergranular fractures:	6.0%

51. These values cannot be compared directly with the average deviations discussed in connection with the static tests. They do not refer to a single specimen or to the mean of a group of specimens. They are values arrived at by subjecting a group of specimens to shots above and below the limit. It would be very pleasant to have a number of limit energies for each material, so that a mean could be formed, but the labor involved precluded this possibility.

Discussion of the Data

52. Since we have data for specimens of 3 sizes from each of 9 treatments a great many types of comparisons can be carried out. The general plan we shall adopt is as follows:

- (a) General survey of size effects in the 9 treatments, including comparisons with typical results for other armor steels.
- (b) Comparison of static and ballistic test results.
- (c) Consideration of precrack and postcrack work in the as-received material.
- (d) Effects of the following variables:

Grain size.
Hardness.
Temper brittleness.

Temperature of isothermal transformation.
Specimen orientation.

Size Effect

53. Referring to Table 3 and Figures 11-13, we note that work density falls off with size quite rapidly as we pass from the 3/8" specimen to the 1-1/8" specimen. In passing on to the 1-7/8" specimen the decrease is slower. This behavior is typical of all the metals we have studied.

54. The magnitude of the size effect ratio for 3/8" and 1-7/8" specimens is well in line with values for other steels, for all the conditions except the temper brittle steel E. The following tabulation shows this.

Material	Microstructure	Hardness	Size Effect Ratio; OX specimens. (Comparison of smallest and largest specimens)
Plate 52151:			
As received	Tempered Martensite	230-234	1.90 - 1.93
B, fine grain	"	237	1.89
D, coarse grain	"	226	1.80
E, coarse grain	"	233	2.92
Bethlehem 4A443A1 (18" plate)	"	238	1.88
Plate 75907, 2"	"	248	1.85
Plate 52151:			
H	400° product	231	2.00
K	575° product	227	1.86
M	625° product	215	2.06
Bethlehem 4A443A1 (18" plate)	Tempered bainite	211	2.43
Carnegie-Ill. 00296	Tempered martensite	246	2.02
Same retreated	Coarse pearlite	202	1.80

55. The temper-brittle treatment E yields the highest (worst) size-effect ratio so far encountered in metals. (A brittle plastic, lucite, yielded values ranging from 3.2 to 3.7 in this type of bend test, using the same specimen sizes employed in our studies of steel.) Looking aside from this exceptional case, we see from the tabulation that the size

effect ratio does not change greatly in passing from tempered martensite through the bainites to coarse pearlite. Aside from treatment E, the range of numbers in the tabulation is only 1/3 of the lowest number. It is clear from these results that the size effect measured by the static bend test is not sensitive to grain size and microstructure. On the other hand, extreme temper embrittlement produced a marked increase of size effect in this steel. We know of one other Ni Cr steel in which the same is true, but the static test cannot be relied on to reveal moderate temper brittleness. The evidence for this statement lies in a study of an 18" Bethlehem plate 4A443A1 (Refs. 2 and 10). Slices sawed from this plate were retempered and slow-cooled. They actually showed an improvement in the slow bend test, but considerable decrease of work value in the ballistic bend test, as compared with the material in the as-received condition.

56. While it was originally hoped that the test would show direct correlation with ballistic penetration data, the above facts and others in our possession make it clear that it cannot do so because of speed influences and radical differences between the patterns of stress and strain. At any speed, the notch bend test simply gives information about the behavior of the metal under a relatively simple pattern of stress and strain, produced at first by the notch and determined later by the deformed notch and the advancing crack. Size-effect bend tests at ballistic speeds are now in course of execution.

Static and Ballistic Behavior

57. Referring to Figure 14, for specimens in both principal directions in the plate we see a general parallelism of the static results and the high speed results, save in 3 cases. These cases are:

- (a) the temper-embrittled material, treatment E;
- (b) the 575 degree isothermal product K;
- (c) the 625 degree isothermal product represented by blocks M and N.

58. We wish not to understand these exceptional cases. As to treatment E, which showed so large a size effect, the static work densities are about 15% lower than those for the corresponding unembrittled material D. This relatively small change is accompanied by a great alteration of the fracture, which is largely intercrystalline in treatment E. The static and ballistic intercrystalline fractures resemble each other closely, in grain texture, under powers of the stereoscopic microscope up to 80. The grains seen are assumed to be pseudomorphs of austenite grains, -not necessarily the ones established in the last anneal, but any prior austenitic grains which had extraneous and disturbed material at their boundaries. A prominent difference between the static and ballistic fractures lies in the presence of numerous fissures in the latter, perpendicular to the main fracture. Correspondingly, the ballistic work falls to 24% of the value for material D.

This indicates a pronounced speed sensitivity of the grain boundary material in the embrittled treatment E.

59. The results for the 575° product, K, are due to a change of fracture type as speed increases. The slow fractures do not show shiny cleavage facets; the fast fractures are a mixed type with cleavage predominating.

60. The behavior of the 625° product depends on direction:

Specimen Axis	Slow Test	Fast Test
OX	Completely ductile fracture near notch. Partway through the specimen fracture alters to the cleavage-facet type.	Ductile fracture confined to small region near notch. Cleavage facet-type predominates.
OY	Cleavage-facet type predominates.	Cleavage-facet type predominates.

61. In all other cases studied, both the slow and fast tests yield fractures which are completely of the ductile type.

Precrack and Postcrack Work in the As-Received Material (Table 6)

62. The ratio of postcrack work to total work is much the same for both the static tests and the ballistic tests of the subplate S1. This result is in line with what we may surmise from the general parallelism of static and ballistic tests for the more ductile treatments of this steel. However, this experiment was necessary to show that the ratio of crack propagation work and total work is nearly constant in the as-received material, when the speed of test is changed by a tremendous factor, approximately 10^8 . It is fairly clear from the preceding section that the result would be quite different if we determined the postcrack work for the materials K and M, which lie on the borderline between ductile and brittle behavior under these test conditions. In general, as speed increases, we expect brittle behavior to show up first in highly worked material, that is, in the postcrack work. After that further increase of speed may result in a decrease of precrack work.

Grain Size (Table 3.)

63. We compare work densities for the treatments B and D, which

have practically the same tensile strength, though not the same hardness:

<u>Specimens:</u>	<u>% Difference, B minus A</u>
OX: 3/8"	3
9/8"	-2
15/8"	-2
OY: 3/8"	-6
9/8"	-4
15/8"	-10

64. The coarse-grained material is slightly superior. This is rather surprising. Further exploration of the matter, covering a wider range of grain sizes, is desirable.

Effect of Tensile Strength (Table 3.)

65. We compare treatments A, B, and C, all being fine-grained tempered martensite with tensile strengths of 101,000, 110,000, and 117,000 PSI, respectively.

<u>Specimens:</u>	<u>% Diff., A - B</u>	<u>% Diff., A - C</u>
OX: 3/8"	16	26
9/8"	13	26
15/8"	16	26
OY: 3/8"	18	29
9/8"	20	30
15/8"	19	28

66. These facts say, the change of work level is directly proportional to the change of tensile strength, over the narrow range of strengths covered; and the actual percentages do not depend on size.

Properties of Isothermal Transformation Products. (Table 5)

67. First we note that the 575° product K shows work-density values almost identical with those of treatments B or D, (all three having comparable tensile strengths; also K is slightly superior to the as-received material SIX. We have discussed the brittleness of the 625° products in Section IV b; but now we note that the work-density values of this pearlitic treatment and of the 400° product are superior to those obtained in the interior bainite of an 18" Bethlehem plate,

and much inferior to those obtained in the same region of an 18" Carnegie-Illinois plate. The W-values are these:

Size	Plate 52151 400° prod., H.	Plate 52151, 625° prod., M.	Beth. Plate 4A43A1 Tempered bainite	Carn.-Ill. Plate DD-427 Tempered Bainite
3/8"	1819	1972	1636 (0.4")	2675
9/8"	1057	1023	978 (1.25")	1913
5/8"	911	955	673	1339

In spite of the wide differences in the results for bainite from the 18" plates, both of them passed the ballistic test at Dahlgren. This means that tensile tests or static bend tests do not reflect the properties responsible for good or poor penetration properties of the material in the interior of a heavy plate. The patterns of stress involved in the bend test and in the full-scale penetration of the plate interior are entirely different. There may be some correlation between the bend test and back-of-plate spalling. Such a correlation was discussed several years ago by Gensamer and his colleagues at Carnegie Institute of Technology (Ref. 11), working on 2" plate. Present data are not extensive enough to prove or disprove the point in the case of heavy plate.

68. The general similarity of the results for the 625° product, 400° product and the tempered bainite from the Bethlehem plate is interesting. However, we know from Figure 14 that the 625° product is speed-sensitive, while the 400° product is not. No data on the speed-sensitivity of Plates 4A43A1 and DD-427 have been obtained; an experiment covering this point is on our program.

Effect of Re-Treatment on Directional Properties

69. In all our studies of rolled and forged Ni Cr plates, the work-density drops when the direction of the specimen axis is changed from parallel to transverse, and from transverse to perpendicular. (There is one slight exception, for plate 4A43A1.) The new data on plate 52151 enable us to study interesting questions:

- (a) To what extent can complete retreatment wipe-out directional variation in rupture-work?
- (b) How does this wipe-out depend on austenitizing temperature?
- (c) How does it depend on the tensile strength after temper?
- (d) How does it depend on microstructure (comparison of tempered martensite and tempered pearlitic steel)?
- (e) Which direction improves the most?

70. The raw material for our answers will now be tabulated:

Treatment. Per cent drop due to change of specimen axis from OX to OY direction.

		Values for 3 sizes			Average
As-received, S1		25,	25,	34	28
As-received, S2		28,	23,	32	28
850°C.	A 120 RPSI	13,	15,	20	16
	B 110	15,	19,	21	19
	C 100	16,	19,	21	18
1100°C.	D 110	7,	20,	17	15
	E* 110 (Temper-brittle)	11,	20,	7	13
M - N		25,	25,	33	28

* Specimen-variations are large for treatment E.

71. Answering the questions posed,

(a) Complete retreatment wiped out about 1/3 of the directional differences in rupture-work of time-quenched and tempered products, the rolling reduction being 91%.

(b) As expected, the data indicate that higher austenitizing temperature gave slightly better wipe-out. This is seen by comparing B with D.

(c) The data do not reveal any definite dependence of this equalization on the tensile strength, when the strength is established by tempering after a given austenitizing treatment. (Compare A, B, C).

(d) No equalization was obtained by retreating to a coarse pearlite.

(e) Comparing as-received material with treatments B and D, (all three being tempered martensites of the same tensile strength), we learn that the work absorption of OX-specimens was not appreciably altered by complete re-treatment. In all fairness, it must be noted that the time-quench applied to the as-received material was longer than that given to materials A to E; also the as-received material was quenched cold, while A, B, C, D, and E were air cooled after the time-quench. Nevertheless, we believe the equalization is mainly due to a genuine improvement in the rupture work of transverse specimens.

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MODELS AND EQUATIONS DESCRIBING THE SIZE EFFECT

Two-Constant Equation for Size Effect

72. In the past many equations have been proposed to describe the impact work or the static bend work as a function of the specimen dimensions. Most of these have been empirical and their application to the data has led to rather poor agreement. Such equations are useful only for the rough design of specimens, or for rough extrapolation to specimen sizes outside the domain covered by experiment. One of the simplest equations for the rupture work of geometrically similar specimens is

$$E = Ad^3 + Bd^2, \quad (4)$$

containing only two constants, A and B. This is equivalent to saying that the work density is

$$W = E/bd^2 = A(d/b) + B/b. \quad (5)$$

73. The idea behind these equations is that the term Ad^3 , proportional to volume, substantially represents the work done before the cracks commence, while the other term, proportional to the area of the cross-section, represents work done in a limited volume close to the final fracture surface. This idea is an over-simplification, but regardless of its merits or its faults, the formulae above deserve to be tested by means of our accurate slow bend data. This has been done for 5 steels, covering a total of 12 heat treatments. Since data for 3 sizes are available, we can determine A and B by least squares, and then can compare W-values calculated from equation (5) with the observed values. For the first 4 steels studied, the discrepancies run from 0 to 13%, and average 3.6%. The fifth steel was plate 52151, and data for a number of specimen groups are given in Table 8. Here the discrepancies between observed and computed values of work range from 0 to 6%, and average 1.7%, which must be considered a very good fit. For infinite size, equation (5) predicts a work density of Ad/b , or 0.8A. Comparing the values of 0.8A with the observed work density values for our largest specimens, the implication is that for this specimen size we have reached work densities only 20 to 30% above those which would apply to very large pieces, provided the same microstructures could be obtained.

Alternative Equation for the Total Work, Precrack Work and Postcrack Work, in Bend Specimens.

74. Some years ago, Southwell (Ref. 12) presented evidence that the work per unit area for geometrically similar impact specimens can be represented in the form

$$E/A = GA^n \quad (6)$$

where G is a constant, n is about 1/4, and A is the area. He applied

this equation to the total work for a cylindrical specimen notched all around, which he employed at that time. We find that it works well for steel specimens showing ductile fracture. The deviations from observed values average about 5%. The same type of formula applies to E_1/A , and E_2/A . As far as ability to represent the data is concerned, there is little choice between equations (5) and (6). The latter has at present no simple physical interpretation. It is useful for rough estimation, since it turns out that E/A is a slowly varying function of A . We give a few typical values of the constants C and A , computed by Mr. W. J. Ferguson, in Table 9. C is the work per unit area when A is one square inch. While ductile-fracture work per unit area (E_2/A) runs as high as 600 ft. lbs./sq.in. in this table, values as low as 250 have occurred in other Ni Cr steels, at a hardness of 270 BHN. Only a few values of E_2/A for brittle fracture of steel 52151 have been obtained. They are of the order of magnitude of 50 ft. lbs./sq. in., and there is some indication that E_2/A drops with increasing size, contrary to the situation in the ductile-fracture case.

Stress Raisers and Mutual Influence of Grains as Causes of Size Effect

75. General Remarks - The theory which we have found most useful in describing size effects of various kinds is based on the universal presence of stress-raisers and regions of structural weakness. For very small specimens, the lack of effective mutual grain-support must also be considered. No matter what the nature of the guilty inhomogeneities may be, we refer to them as flaws. The word is meant to embrace cavities, inclusions, precipitates, regions of chemical segregation, regions in which a number of grains are unfavorably oriented for slip, and regions of high self-balancing stresses, on either the macroscopic or the microscopic scale. To see how flaws of any kind whatever give rise to size effects, we only need to consider a few simple facts. In any material there are flaws which are serious and others which are relatively harmless. It must not be supposed that the severity of a flaw depends on its size. In some cases there is such a dependence, and in other cases the contrary is true. For example, it is generally believed that in aging phenomena the maximum loss of ductility is associated with a critical size of the precipitate particles. As another extreme example, we note that all spherical blowholes in an ideal elastic medium produce nearly the same stress concentration, so long as they are well separated. Because of these facts we prefer to classify the stress-raisers according to the severity of the stress concentration they produce, and the structure-weakeners according to the decrease they cause in the value of the property which happens to interest us. Since the general effects of stress concentration and of structural weakness are similar, it is ordinarily reasonable and proper to treat both classes of entities on a common basis, so this will be done in the present discussion. It conduces to simplicity if we focus attention on the more serious classes of flaws, those which bear the major responsibility for determining the value of some mechanical property of the material, or for causing fracture. There are now two extreme cases to consider.

76. First, it may occur that the serious flaws are so infrequent that there is only a small chance of having one of them in a specimen chosen at random. In this case, it is clear that an increase of specimen size leads to an increased expectation of having a flaw in the specimen. The average strength and the average work absorbed would therefore fall off if the specimen size were increased. In such a case the metal would be described as very good on the average and the size effect would be large. In the second extreme case, we suppose that the serious flaws are so numerous that every specimen is practically certain to contain one or more of them. Then the metal is very poor, but the size effect is small. Intermediate cases can be discussed in similar vein.

77. The situation becomes more complicated when several coexisting classes of flaws are present. For example it would be possible to have one class so numerous that it exerts its entire influence in causing size effect at very low absolute size levels, and another class, more serious but much less frequent, which comes into effective play only at high levels of size. In this case it is possible to choose a specimen of such dimensions that the size effect due to the first class is complete, or saturated, as we may say, while the size effect due to the other need not be considered. Further complications arise when the cooperative action of flaws must be taken into account, but in all cases the general ideas which clarify the situation for us are the same.

78. In the above we have had in mind specimens of such dimensions that mutual support of the grains is substantially complete. In short we have neglected the weakness characteristic of single crystals, or of specimens which have a relatively small number of grains in the dangerous cross section. Borrowing a terminology used in other fields, we shall call such specimens plural crystals, to distinguish them from the polycrystalline specimens employed in ordinary engineering practice. To be more definite we propose that a specimen should be called plural-crystalline when it contains more than one grain and less than the number of grains required to confer upon it 95% of the maximum property-value obtainable by variation of the specimen size. This definition, though arbitrary, is likely to appeal to test engineers. It is clear that size effect is operative in the plural-crystalline range of sizes, and that the occurrence of maximum properties is due to a competition between the development of mutual grain-support and the weakening effects of the flaws. From evidence which need not be quoted here, it is clear that in metals of customary grain-size range, the boundary between the plural crystal and the polycrystal is of the order of magnitude of 10^{-2} inches.

Simplified Flaw Theory

79. The above remarks have been included to explain the general nature of the size effects studied in this report. The mathematics of the subject, begun by Weibull (Ref. 8), has been developed further during the war years by Rosen and by Ruark (Ref. 9). As the theory

stands, it refers primarily to brittle fracture. Rosen's analysis is general, starting from the bare assumption that there is a chance

$$F(S) \quad dv \quad dS,$$

such that the volume element dv contains a flaw which will cause fracture of the specimen at a stress between S and $S + dS$. From a curve showing the distribution of breaking strengths for a large number of specimens, it is possible to find the flaw-probability function F . It is also possible to find it from a curve showing the average strength as a function of specimen size. Conversely, knowing the function F , it is possible to write the equation of this size-effect curve. Unfortunately, neither the present data nor any other size effect data known to us are sufficiently extensive to yield a reasonably accurate determination of the function F . For this purpose we would need to study at least 20 specimens at each of 10 sizes; and many times that number would be still better.

80. The essential features of all such theories can be seen from one simple example, dealing with the size effect of the rupture work of brittle specimens, (Ref. 9 and Ref. 2). Let us consider the consequences of the following assumptions:

(a) There is just one class of flaws, all of which produce the same stress concentration.

(b) In the absence of a flaw the rupture work is the normal value E_n . If the specimen contains one or more flaws, it breaks with a lower value, E_f .

(c) The average number of flaws in a unit volume is f and the volume exposed to risk of fracture is V .

81. Now, a standard formula from the theory of probability tells us that the chance of having no flaw at all in a specimen with the "dangerous volume" V is

$$P_0 = e^{-fv}. \quad (7)$$

But the volume exposed to risk of fracture is unknown in notched specimens. To avoid this difficulty we make another assumption, that in geometrically similar specimens the volume exposed to risk is proportional to specimen size. It is not claimed that this is reasonable in all cases, but when it is true we can make use of a conventional volume V' , where

$$V' = NV,$$

and N is simply an unknown coefficient of proportionality. Writing

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$f' = f/N$, we can put equation (7) in the form

$$p_o = e^{-f'V'} \quad (8)$$

In fact, all that we can determine from this argument is the product fV or $f'V'$. The fraction of the specimens breaking with work E_n is $e^{-f'V'}$; the fraction with work E_f is $1 - e^{-f'V'}$. Therefore the average work-density for a group with conventional volume V' is

$$E = E_n e^{-f'V'} + E_f (1 - e^{-f'V'})$$

or;

$$E = E_f + (E_n - E_f) e^{-f'V'} \quad (9)$$

This is the equation of the size effect curve, under these oversimplified conditions. If we calculate the standard deviation of the rupture work as a function of fV and express it as a percentage p of the mean rupture work, we find that it is zero for very small or very large specimens and that it passes through a maximum when V is so chosen that $fV = 0.7$. This is easy to understand: the small specimens contain practically no flaws; the large ones, saturated with flaws, nearly all break with the rupture work E_f .

82. We regret our inability to apply the general flaw theory numerically to our present data, but thought it best to include these explanations so that the possibilities for future advance can be clearly seen. It must be pointed out that all we can get from any theory of this kind is the function F , or a related quantity such as fV . If V can be inferred or measured, the analysis of the mechanical data can give us the spacing of the flaws of a given class (through a knowledge of the number per unit volume) but it cannot tell us how large they are, or anything else about them.

SUMMARY AND RECOMMENDATIONS

83. Nine treatments of a 2.5" Carnegie-Illinois nickel chrome steel plate, aluminum-killed, have been subjected to bend tests at quasi-static speed. Three sizes of specimens, all geometrically similar, were employed. Bend tests at ballistic speed were also carried out, using a single size, and one treatment of this steel was tested in such a way as to determine the precracking work and the postcracking work.

84. Table 1 describes the materials. Slow-bend test results are recorded as values of a conventional work-per-unit-volume, W , required to produce fracture. Ballistic-bend results are recorded by giving the bullet energy required to produce fracture under our experimental conditions. The principal results are these:

(a) Size effects in this material are well in line with those for other armor steels subjected to the same low-speed bend test. In

the range of sizes and treatments covered, the size effect of steel 52151 is quite insensitive to grain size and microstructure. Changes of tensile strength from 100 to about 117 KIPS do not change it. There is no important change of size effect in passing from tempered martensitic material to products transformed isothermally at 400, 575 and 625°C. However, a large increase in size effect was observed when a certain treatment (D) was strongly embrittled by retempering and slow cooling. The resultant product was so brittle that the fracture was largely intercrystalline at room temperature.

(b) Comparison of static and ballistic behavior. There is general parallelism of the static and ballistic results except in the cases of the temper-embrittled material (treatment E), and isothermal products transformed at 575° (treatment K) and at 625° (treatment M, N). In these three cases, the dynamic work falls far below expectations based on the static results. The result for the temper-embrittled material points to pronounced speed-sensitivity of grain-boundary material involved in fracture. In the case of the isothermal products we are dealing with a speed-induced change from ductile-type fracture to cleavage-type, both being transgranular.

(c) The results on postcracking work in the as-received material (treatment S) are best expressed by saying that the ratio of postcracking work to total work is about the same at both low and high speeds. However, this material yielded ductile fractures at both speeds. We believe that if the test were repeated on one of the treatments which gave low ballistic-bend results (E, or K, or M-N), we would find that the postcracking work would decrease before the precracking work, as speed is increased.

(d) Effects of treatment on absolute level of the work density - While the size effect was insensitive to most of the treatment-variations, as recorded above, the work density itself is sensitive to some of these changes. Grain-size variation made little difference. Increase of tensile strength of a tempered martensitic material by 1% raises the work density by an amount ranging from 1.5 to 2%, over the range of tensile strengths covered in these experiments. The isothermal products formed near 575° gave a work density as high as that for tempered martensites of the same tensile strength, but those formed at much higher or much lower temperatures are inferior in this respect.

(e) Directional Properties - Transverse specimens showed lower work-density values than longitudinal ones in all treatments studied. The effects of complete re-heat treatment on the directional differences are of interest. In the tempered martensitic and tempered bainitic materials, which show ductile fracture, retreatment wipes out one third of the difference in work density. This equalization increases with austenitizing temperature, as might be expected. The equalization arises from improvement of the work density of transverse specimens. There is little or no change in the work density of longitudinal ones. On the other hand, no equalization was detected in material retreated to a coarse pearlitic structure.

(f) Equations describing the size effect - The rupture work

can be closely expressed by the simple empirical equation

$$E = Ad^3 + Bd^2$$

where A and B are constants for the particular material and treatment employed, and d is a typical dimension of the specimen. Another equation of about the same accuracy, is

$$E/A = CA^n,$$

where C is a constant and n is an exponent having a value of about 1/4.

(g) Explanations of size effect - The size effect is mainly due to the influence of flaws and to grain size-effects, the former predominating in the range of specimen sizes customarily used. The arguments for this view, outlined in earlier reports, are restated in this report for the convenience of the reader. The fact that the size of the advancing crack-head region does not scale up with specimen size is an accessory cause of size effect.

(h) The results suggest desirable extensions of the work. Additional size effect studies should be made at ballistic speed, using this well-studied steel. They should cover a very good martensitic material (treatment A or B) the temper-brittle material (treatment E), and one or more of the isothermal products. Facilities for these experiments are now practically complete. They may be expected to yield a fair idea as to the speed-dependence of size-effect in nickel chrome armor steel. The experiments reported here show that strain rate is a significant factor in determining the rupture work of the more brittle treatments. It now becomes important to ascertain the approximate strain rates which bring about this decrease of rupture work. We do not know whether the effect is sharply localized in the strain-rate scale or whether, as seems more likely, the deterioration is a gradual one as speed increases. A technique for studying this matter by plate-penetration tests at various velocities has been worked out at NRL and has been applied to aluminum alloys successfully, (Ref. 13). Similar work should be done on steel, using material from Plate 52151. It is believed that such penetration tests may be more rewarding than further study of rupture work in bend tests at various speeds, although the latter are also desirable.

85. Acknowledgments - We appreciate the courteous cooperation of Carnegie-Illinois officials mentioned in this report. The Special Alloys Section, and the Metallographic Section, Division of Metallurgy, NRL, have been helpful in discussing the microstructures found in the steel employed.

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SUMMARY OF NRL REPORT O-2412

(This material is reproduced here to give the reader a quick review of all previous size effect studies carried out at the University of North Carolina under NRL contracts for BuOrd, prior to work on Plate 52151)

1. In full scale ballistic tests, the onset of brittle fracture occurs at a hardness value much below that at which the same effect occurs in small-scale laboratory tests, carried out on a model basis. This lack of similitude, called size effect, appears to be of the same nature as similar effects encountered in other mechanical tests, notably in those carried out with notched specimens.
2. In the absence of sufficient data on size effects and the factors which determine their magnitude and importance, we have made experiments on several NiCr steels, and on 24ST aluminum. Using geometrically similar specimens, we have made tensile tests and static bend tests with precision equipment designed and constructed for the purpose. The equipment is fully described.
3. In specimens of customary size, we are not concerned with the rise of strength which occurs as one passes from a single crystal to the polycrystalline condition. We study only specimen sizes which are of practical interest. In both tensile and bend tests, it is found that no appreciable size effect is encountered during the period of plastic flow. The size effect comes in through the earlier onset of cracking in the larger specimens, and the reduction of the work, per unit volume of specimen, required to propagate the main crack.
4. In the tensile test, the most prominent size effect is found in the reduction of area, which becomes smaller as the size increases. This loss of ductility is not new, but in nearly all or perhaps all, previously recorded cases, the effect is complicated by other factors such as inhomogeneity of the material, or differences in the rolling reduction of the various sizes tested. Correlated with this change of area-reduction, there is a slight decrease of the maximum true stress for large specimens.
5. In bend tests, the details of the size effect can be followed much more easily. The quantities dealt with are the energy required to carry the specimen up to maximum load, the energy required to finish the rupture and the total energy consumed. To reduce the results to a comparable basis, each of these energies is divided by a quantity proportional to the cube of a typical specimen dimension. Thus we deal with work densities, W_1 , W_2 , and W , corresponding to the three energies just mentioned. If there were no size effects these quantities would be independent of specimen size. Things being as they are, all three quantities drop as size increases, and the ratio of W -values for two specimens of different dimensions is a good measure of the effects to be studied.

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6. It is found that the larger the specimen, the earlier the small cracks are formed in the notch, but this is not the only deleterious effect of large size. The fraction of the work done after the cracks are formed is usually less for the larger specimens. Those facts can be explained by noting (1) that the chance of having a serious stress-raiser in the neighborhood of the notch increases with specimen size, and (2) that after the main crack is formed it constitutes a sharper notch in a large specimen than in a small one, at any given stage of the test. This is true because the effective radius of the crack head does not rise when the size of the specimen is increased.

7. We now consider the evidence gathered as to the effect of metallurgical and testing variables on the magnitude of the size effect in bending. It is not possible to consider all interesting variables because the material could not be chosen in accordance with ideal requirements.

(a) Microstructure. The size effect seems quite constant for tempered martensites, all within a rather narrow hardness range, running from 238 to 265, BHN. It is greater for tempered bainites. Special tests were made on 18" Class B armor material, treated to make it temper brittle and also treated to produce some ferrite and pearlite. Controls were run on the same material in the as-received condition. The bend test, and also Izod impact tests, failed to reveal temper brittleness in this steel, but its existence has been demonstrated by special ballistic tests at NRL. Both air-cooling and furnace-cooling lower the work-density values and raise the size effect, in comparison with the results for tempered martensite from the same plate.

(b) Inclusion. Materials used for inclusion-counts were supplied by the Special Alloys Section, NRL. For each specimen, the total inclusions and the number of sharp or angular ones, per square inch of etched surface, had been determined. The bend tests showed larger size effect with increase of total inclusions and of angular ones. The separate effects of these two classes of stress-raisers could not be determined from the specimens available, and this matter requires further study.

(c) Hardness. Similar work at Carnegie Institute of Technology showed that the size effect in an armor steel increased with hardness. In our work, variations of composition from one material to another have prevented us from adding quantitative information on this point.

(d) Orientation. Use of material from the very uniform center portion of an 18" plate permitted us to work with specimens with their axes in all three principal orientations, relative to the length-direction of the plate. These studies, and similar ones on another steel, usually show decrease of work-density and increase of size effect as we pass from longitudinal to transverse, and from transverse to perpendicular specimens. Similar effects are found in rolled 24ST aluminum, where they are very pronounced.

(e) Speed of deformation. In our work testing speeds are low and are so controlled that the influence of speed can be neglected. It is pointed out, however, that speed effects are by no means negligible when one passes from the conditions of static tests to those of tests at Izod-speed or bullet-speed. The changes of mechanical properties encountered at high speeds are complex and will require much further study, on armor materials, before it will be possible to understand completely the loss of penetration-resistance encountered in full-scale ballistic tests, as compared with small-scale tests. This loss is evidently due to size effects and speed effects, acting together. Attention is directed to the fact that in the rapid deformation of metals plastic flow and cracking occur at elevated temperatures, so that the properties of armor materials at such temperatures should be considered.

8. Theoretical studies have been carried on to help us understand the influence of inclusions; or other stress-raisers. In the simplest case, that in which a single class of flaws is important, the work-density in geometrically similar specimens is related to a typical specimen dimension L according to the equation

$$W = W_f - (W_n - W_f) \exp(-fL)$$

where W_f is the lowest work-density encountered and W_n the highest one; f is the average number of flaws available to cause fracture in a specimen for which L is equal to unity. There is reason to believe that this equation gives the general trend of affairs for cases encountered in practice. It predicts relatively small changes in work-density for sufficiently high values of fL . This type of behavior is clearly shown in our measurements on 24ST aluminum. In all the steels work-density is still decreasing rapidly at the top of our available size-range. The influence of the arrangement of stress raisers is believed to be very important, because of their cooperative influence. For example, if two spherical cavities are brought close together the stress concentration rises without limit, starting at the value 2 pertaining to an isolated sphere. The variations we have encountered in our measurements of size effect in relatively clean and relatively dirty steels are not as great as we might have expected. Part of the explanation may lie in the ability of a ductile metal to protect itself from the bad effects of stress-raisers by plastic flow. We are inclined to believe, however, that the influence of the general pattern of stress-raisers is the most potent factor available for explaining otherwise incomprehensible variations in the behavior of different heats of steel.

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HISTORY OF 2.5" PLATE 52151, WITH ANALYSIS, HEAT TREATMENT AND PHYSICAL PROPERTIES. FROM CARNEGIE-ILLINOIS STEEL COMPANY.

Location of Pieces in Rolled Plate

The as-rolled plate, 150" x 100" x 2.5", was from top slab cut of ingot (See Ingot and Slab Sizes) Figure 7 shows general lay-out of all pieces on rolled plate. Piece 65" wide and 108" long (indicated by dotted lines) treated in one piece. Plates 52151S1 and 52151S2 flame cut out after treating as shown. All plates were stamped with numbers and letters on top end as indicated on sketch; in addition, plates were stamped to show rolling direction and top and bottom with reference to ingot position.

Open Hearth Record. Heat No. 18L153. Date Cast. April 9, 1944

Ordered Analysis

<u>C</u>	<u>Mn</u>	<u>P</u>	<u>S</u>	<u>Si</u>	<u>Ni</u>	<u>Cr</u>
.26/.28	.20/1.30	.025 max.	.035 max.	.07/0.08	3.40/3.60	1.40/1.60

2.5 lb. aluminum per ton in ladle.

Charge

Mi low-in Hot Metal.....	62000%
Mi Cold Metal.....	6000%
Mi Scrap.....	88000%
	156000%
Lime.....	6000%
Cro.....	None

Melt

1.42 C @ 9:30 P.M. - April 8, 1944.

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Furnace Additions

Cinder..... 3200#
 Lime..... 4000#
 Ni..... 3712# @ 1:00 A.M.
 Cr..... 3700# @ 2:06 A.M.
 Mn..... 75# @ 2:06 A.M.

Ladle Additions

Coal..... 75#
 Fe-Si (50%)..... 225#
 Al..... 190#

Tap

2:20 A.M., April 9, 1944 - Spike Time - 15 Minutes.

<u>C</u>	<u>Mn</u>	<u>FeO</u> (Slag)
.14/.15	.16/.17	8.21

Tapping Time..... 3 minutes, steel - 3 minutes, slag
 Held..... 4 minutes
 Pouring Time..... 20 minutes

Temperatures

Bath..... 2945°F @ 1:45 A.M.
 Tap..... 2980°F
 Start Pouring... 2875°F
 Finish Pouring.. 2830°F

Nozzle

2-1/2"

Time of Heat

13-1/4 hours
 Working Time - 4-1/2 hours

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Pouring (Ingot to make plate 52151 only)

No. 1 Ingot, - 28" x 60", @ 32000#.
 Time to Pour Body, 1 minute, 30 seconds.
 Total Time, 1 minute, 55 seconds.
 No pouring irregularities.
 No aluminum fed in mould.

Chemical Analysis

Heat 18L153	C	Mn	P	S	Si	Ni	Cr
Ladle	.29	.22	.013	.020	.08	3.59	1.66
52151S1(Check)	.27	.22	.014	.020	.07	3.62	1.71
52151S2(Check)	.27	.22	.014	.020	.08	3.60	1.72

The check analyses shown above were run on the test bars for each of the respective plates (See Figure 7) No other check analysis was taken.

Ingot and Slab Sizes

Two slabs cut from ingot No. 1:
 No. 2 cut - 55" x 11" x 73 1/2" @ 12000#
 No. 3 cut - 55" x 11" x 69 1/2" @ 11290#
 TOTAL 23290#

Ingot to Slab Practice - 72.9%

No. 2 cut (52151) rolled to plate 150" x 100" x 2 1/2".
 This is top cut. No. 1 cut is discard.

Heat Treatments and Physical Properties. See Table 1.

Dates

With a view to future aging studies, we record the following:

Cast, 9 April, 1944
 Main body shipped to NRL About 20 July, 1944
 11 Blocks, (A to F, and H to N),
 shipped to NRL About 13 Oct., 1944
 Size effect tests on main body:
 Size effect tests on blocks A to N:

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TABLE I. HEAT TREATMENT AND PHYSICAL PROPERTIES OF SUB-PLATES CUT FROM PLATE 52151.

Product	Temp., °C	Maintain	Cool	Plates	Y.P.	T.S.	% Elong.	% R.A.
Fine-grained Temp. Mart.	835	1-1/4 hrs.	Water quench cold in 3-1/2 minutes.	S1	91910	110500	24.0	67.7
	633	4 hrs.	Water quench cold.	S2	94560	111350	25.0	70.8
Fine-grained Temp. Mart.	850	1-1/4 hrs.	Water quench 140 sec. Air cool cold.					
	660	16 hrs.	Water quench cold.	A	82880	101300	29.5	71.8
	655	4 hrs.	Water quench cold.	B	92120	110500	26.5	71.4
	630	4 hrs.	Water quench cold.	C	97640	117100	23.0	70.6
Coarse-grained Temp. Mart.	1100	1-1/4 hrs.	Water quench 175 sec. Air cool cold.					
	655	4 hrs.	Water quench cold	D	85540	109300	26.5	70.9
Temper Brittle	655	4 hrs.	Air cool cold.	E	89420	112100	25.0	68.9
	500	8 hrs.	Furnace cool.					

NOTE: Plates 52151S1 and 52151S2 treated in large furnace at Homestead. Balance of plates treated in small tool treating furnace and lead bath at South Charleston.
S1 and S2 are called the "as-received condition" in this report.
Desired tensile strengths (PSI) were as follows: A, 100000; B, 110000; C, 120000; D, 110000; E, 110000

(1)

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ISOTHERMAL TREATMENTS

TABLE 1, Continued.

Product	Temp., ° C.	Maintain	Cool	Plate	Y.P.	T.S.	% Elong.	% R.A.
Partially transformed to 400° product.	1100	1-1/4 hrs.	Trans. to lead bath.	H	89860	111000	25.0	66.0
	400	1 hr.	Water quench cold.	J	91740	111500	23.5	65.8
	630	4 hrs.	Water quench cold.					
Completely transformed to 575° product.	1100	1-1/4 hrs.	Trans. to lead bath.	K	77540	112500	24.0	67.5
	575	30 hrs.	Water quench cold.	L	75340	111800	24.5	67.4
	630	2 hrs.	Water quench cold.					
Completely transformed to 625° product	1100	1-1/4 hrs.	Trans. to lead bath.	M	71600	106200	27.0	66.1
	625	8 hrs.	Water quench cold.	N	62000	108000	24.5	64.7
	575	2 hrs.	Water quench cold.					

Desired tensile strength, 110,000 PSI for H, J, K, L; 105,000 PSI for M and N.

(2)

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TABLE 2 DETAILED RESULTS OF STATIC BEND TESTS

Carnegie Illinois Plate 52151, Plates A to F, and S 2 (for comparison with the condition prior to special heat treatments. 3/8" OX Specimens. Work densities in ft. lbs./cu.in.)

Spec. No.	W ₁	W ₂	W
A7X	1553	1194	2747
A8X	1496	1224	2720
A9X	1497	1282	2779
A10X	1569	1175	2744
Means	1529	1219	2748 $\pm$ 9
B7X	1259	1015	2274
B8X	1342	1060	2402
B9X	1224	1138	2362
B10X	1308	885	2193
Means	1283	1025	2308 $\pm$ 43
C7X	1167	929	2096
C8X	1154	935	2089
C9X	1096	875	1971
C10X	1017	860	1877
C11X	1143	903	2046
C12X	1031	1028	2059
C13X	1016	981	1997
C14X	1161	929	2090
C15X	1080	910	1990
C17X	1086	1018	2104
C18X	1037	1001	2038
C19X	1061	957	2018
C20X	1044	982	2026
C21X	974	1045	2019
C22X	1213	880	2093
C23X	1032	1048	2080
Means	1082	955	2037 $\pm$ 1.6
D7X	1228	1077	2305
D8X	1199	1036	2235
D9X	1165	999	2164
D10X	1183	1077	2260
Means	1194	1047	2241 $\pm$ 24
E7X	1051	944	1995
E8X	1017	776	1793
E9X	1058	912	1970
F10V	1107	943	2050
Means	1058	894	1952 $\pm$ 46

TABLE 2 (Cont'd)

Spec. No.	W ₁	W ₂	W
S2X7	1172	1317	2489
S2X8	1101	1126	2227
S2X9	1329	1045	2374
S2X10	1215	1060	2275
Means	1204	1137	2341 ± 52

3/8 in. OY Specimens. Plates A to E, and S 2

A7Y	1439	1009	2448
A8Y	1398	991	2389
A9Y	1402	982	2384
A10Y	1401	985	2386
Means	1410	992	2402 ± 13
B7Y	1176	870	2046
B8Y	1148	870	2018
B9Y	1084	741	1825
B10Y	1222	764	1986
Means	1158	811	1969 ± 42
C7Y	964	781	1745
C8Y	997	721	1718
C9Y	973	752	1725
C10Y	935	741	1676
Means	967	749	1716 ± 14
D7Y	1226	936	2162
D8Y	1222	865	2087
D9Y	1167	820	1987
D10Y	1279	799	2078
Means	1222	855	2077 ± 27
E7Y	914	764	1678
E8Y	988	670	1658
E9Y	1024	838	1862
E10Y	1004	723	1727
Means	982	749	1731 ± 38
S2Y7	966	865	1831
S2Y8	801	877	1678
S2Y9	837	782	1619
S2Y10	839	775	1614
Means	861	825	1686 ± 42

TABLE 2 (Cont'd)

3/8 in. Specimens. Plates H, K, M, N and S 1

Spec. No.	W ₁	W ₂	W
OX SPECIMENS			
H7X	854	1101	1955
H8X	824	991	1815
H9X	750	966	1716
H10X	742	1044	1786
Means	793	1026	1819 ± 42
K7X	1279	1044	2323
K8X	1364	1057	2421
K9X	1230	1054	2284
K10X	1365	1090	2455
Means	1309	1061	2370 ± 39
M7X	1017	916	1933
M8X	1072	999	2071
M9X	1042	891	1933
M10X	1118	8833	1951
Means	1062	910	1972 ± 29
S1X7	1427	1043	2470
S1X8	1282	1275	2557
S1X9	982	1063	2045
S1X10	963	1290	2253
Means	1164	1168	2332 ± 105
OY SPECIMENS			
N7Y	991	601	1592
N8Y	985	497	1482
N9Y	790	448	1238
N10Y	889	727	1616
Means	914	568	1482 ± 71
S1Y7	1032	789	1821
S1Y8	1000	823	1823
S1Y9	856	766	1622
S1Y10	904	793	1697
Means	948	793	1741 ± 47

TABLE 2 (Cont'd)

1 1/8 in. OX Specimens. Plates A to E, and S 2

Spec. No.	W ₁	W ₂	W
A4X	1045	650	1695
A5X	1035	678	1713
A6X	1016	685	1701
Means	1032	671	1703 ± 5
B4X	945	618	1563
B5X	746	637	1383
B6X	779	712	1491
Means	823	656	1479 ± 45
C4X	643	584	1227
C5X	683	566	1249
C6X	692	607	1299
Means	673	585	1258 ± 19
D4X	842	678	1520
D5X	825	662	1487
D6X	830	707	1537
Means	832	682	1514 ± 13
E4X	651	334	985
E5X	668	311	979
E6X	640	178	818
Means	653	274	927 ± 51
S2X4	778	579	1357
S2X5	778	569	1347
S2X6	788	592	1380
Means	782	580	1362 ± 9
A4Y	865	570	1435
A5Y	902	568	1470
A6Y	911	545	1456
Means	893	561	1454 ± 9
B4Y	711	473	1184
B5Y	654	526	1180
B6Y	664	476	1140
Means	676	492	1168 ± 13
C4Y	610	432	1042
C5Y	562	431	993
C6Y	570	446	1016
Means	581	437	1018 ± 12
D4Y	739	505	1244
D5Y	708	541	1249
D6Y	600	543	1143
Means	682	530	1212 ± 32
E4Y	588	130	718
E5Y	591	151	742
E6Y	606	157	763
Means	595	146	741 ± 11
S2Y4	620	416	1036
S2Y5	589	436	1025
S2Y6	644	432	1076
Means	618	428	1046 ± 14

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TABLE 2 (Cont'd)

1 1/8 in. OX and OY Specimens. Plates H, K, M, N, and S 1

Spec. No.	W ₁	W ₂	W
H4X	552	566	1118
H5X	569	519	1088
H6X	388	576	964
Means	503	554	1057 $\pm$ 44
K4X	825	615	1440
K5X	830	638	1468
K6X	847	639	1486
Means	834	631	1465 $\pm$ 11
M4X	739	541	1280
M5X	704	251	955
M6X	668	165	833
Means	704	319	1023 $\pm$ 122
S1X4	781	622	1403
S1X5	738	631	1369
S1X6	729	640	1369
Means	749	631	1380 $\pm$ 11

OY SPECIMENS

N4Y	649	144	793
N5Y	634	153	787
N6Y	609	116	725
Means	630	138	768 $\pm$ 20
S1Y4	611	414	1025
S1Y5	640	428	1068
S1Y6	589	408	997
Means	613	417	1030 $\pm$ 18

TABLE 2 (Cont'd)

1 7/8 in. Specimens. Plates A to E and S 2

Spec. No.	W ₁	W ₂	W
A1X	883	573	1456
A2X	873	602	1475
A3X	852	581	1433
Means	869	586	1455 ± 10
B1X	677	517	1194
B2X	678	527	1205
B3X	739	517	1256
Means	698	521	1219 ± 18
C1X	572	494	1066
C2X	569	493	1062
C3X	648	451	1099
Means	597	479	1076 ± 11
D1X	681	555	1236
D2X	733	520	1253
D3X	712	535	1247
Means	709	537	1246 ± 4
E1X	540	113	653
E2X	583	124	707
E3X	551	96	647
Means	558	111	669 ± 18
S2X1	748	549	1297
S2X2	692	520	1212
S2X3	701	492	1193
Means	714	520	1234 ± 30
A1Y	697	432	1129
A2Y	791	414	1205
A3Y	754	419	1173
Means	747	421	1168 ± 19
B1Y	590	361	951
B2Y	619	370	989
B3Y	535	372	907
Means	581	367	948 ± 20
C1Y	480	359	839
C2Y	516	333	849
C3Y	508	342	850
Means	501	345	846 ± 3
D1Y	603	412	1015
D2Y	645	380	1025
D3Y	653	404	1057
Means	634	399	1033 ± 12
E1Y	496	122	618
E2Y	501	98	599
E3Y	525	121	646
Means	507	113	620 ± 11
S2Y1	493	340	833
S2Y2	485	344	829
S2Y3	534	336	870
Means	504	340	844 ± 12

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TABLE 2 (Cont'd)

1 7/8 in. OX and OY Specimens. Plates H, K, M, N, and S1

Spec. No.	W ₁	W ₂	W
H1X	445	440	885
H2X	481	475	956
H3X	450	441	891
Means	459	452	911 $\pm$ 21
K1X	690	547	1237
K2X	743	557	1300
K3X	760	524	1284
Means	731	543	1274 $\pm$ 13
M1X	602	448	1050
M2X	644	198	842
M3X	629	345	974
Means	625	330	955 $\pm$ 53
S1X1	680	487	1167
S1X2	737	501	1238
S1X3	713	505	1218
Means	710	497	1207 $\pm$ 19

OY SPECIMENS

N1Y	513	80	593
N2Y	541	114	655
N3Y	573	102	675
Means	542	99	641 $\pm$ 23
S1Y1	446	333	779
S1Y2	445	334	779
S1Y3	483	346	830
Means	458	338	796 $\pm$ 16

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TABLE 3.

SUMMARY OF STATIC BEND TESTS, WITH SIZE EFFECT RATIO;
WORK DENSITIES IN FT.LB/CU. IN.

lock No.	Spec. Size	W ₁	W ₂	W	% Dev. in W	$\frac{W_{0.375}}{W_{1.125}}$	$\frac{W_{1.125}}{W_{1.875}}$	$\frac{W_{0.375}}{W_{1.875}}$	Brinell Hardness Number
AX	0.375	1529	1219	2748	0.3				
	1.125	1032	671	1703	0.3				
	1.875	869	586	1455	0.7	1.61 $\pm$.007	1.17 $\pm$.009	1.89 $\pm$.01	214
BX	0.375	1283	1025	2308	1.9				
	1.125	823	656	1479	3.0				
	1.875	698	521	1219	1.5	1.56 $\pm$.05 ₅	1.21 $\pm$.04	1.89 $\pm$.05	237
CX	0.375	1082	955	2037	0.6				
	1.125	673	585	1258	1.5				
	1.875	597	479	1076	1.0	1.62 $\pm$.02 ₆	1.17 $\pm$.02	1.89 $\pm$.02	252
DX	0.375	1194	1047	2241	1.1				
	1.125	832	682	1514	0.9				
	1.875	709	537	1246	0.3	1.48 $\pm$.02 ₁	1.21 $\pm$.01	1.80 $\pm$.02	226
EX	0.375	1058	894	1952	2.4				
	1.125	653	274	927	5.7				
	1.875	558	111	669	2.7	2.11 $\pm$.13	1.39 $\pm$.09	2.92 $\pm$.10	233
S ₂ X	0.375	1204	1137	2341	2.2				
	1.125	782	580	1362	0.7				
	1.875	714	520	1234	2.4	1.72 $\pm$.04	1.10 $\pm$.03	1.90 $\pm$.06	230
AY	0.375	1410	992	2402	0.5				
	1.125	893	561	1454	0.6				
	1.875	747	421	1168	1.6	1.65 $\pm$.01 ₃	1.24 $\pm$.02	2.06 $\pm$.04	213
BY	0.375	1158	811	1969	2.1				
	1.125	676	492	1168	1.1				
	1.875	581	367	948	2.1	1.68 $\pm$.04	1.23 $\pm$.03	2.08 $\pm$.06	231
CY	0.375	967	749	1716	0.8				
	1.125	581	437	1018	1.2				
	1.875	501	345	846	0.4	1.69 $\pm$.02 ₄	1.20 $\pm$.02	2.03 $\pm$.01	248
DY	0.375	1222	855	2077	1.3				
	1.125	682	530	1212	2.6				
	1.875	634	399	1033	1.2	1.71 $\pm$.04 ₆	1.17 $\pm$.03	2.01 $\pm$.04	215

TABLE 3 (Continued).

lock No.	Spec. Size	W_1	W_2	W	% Dev. in W	$\frac{W_{0.375}}{W_{1.125}}$	$\frac{W_{1.125}}{W_{1.875}}$	$\frac{W_{0.375}}{W_{1.875}}$	Brinell Hardness Number
EY	0.375	982	749	1731	2.2				
	1.125	595	146	741	1.5				
	1.875	507	113	620	1.8	$2.34 \pm .06_3$	$1.20 \pm .03$	$2.79 \pm .08$	228
S	0.375	861	825	1686	2.5				
	1.125	618	428	1046	1.3				
	1.875	504	340	844	1.4	$1.61 \pm .04_5$	$1.25 \pm .02$	$2.00 \pm .06$	232
HX	0.375	793	1026	1819	2.3				
	1.125	503	554	1057	4.2				
	1.875	459	452	911	2.3	1.72	1.16	2.00	231
KX	0.375	1309	1061	2370	1.6				
	1.125	834	631	1465	0.8				
	1.875	731	543	1274	1.0	1.62	1.15	1.86	227
MX	0.375	1061	910	1972	1.5				
	1.125	704	319	1023	11.9				
	1.875	625	330	955	5.5	1.93	1.07	2.06	215
SLX	0.375	1164	1168	2332	4.5				
	1.125	749	631	1380	0.8				
	1.875	710	497	1207	1.6	1.69	1.14	1.93	234
NY	0.375	914	568	1482	4.8				
	1.125	630	138	768	2.6				
	1.875	542	99	641	3.6	1.93	1.20	2.31	219
SLY	0.375	948	793	1741	2.7				
	1.125	613	417	1030	1.7				
	1.875	458	338	796	2.0	1.69	1.29	2.19	237

The per cent deviation in W refers to the average deviation of the mean, not to the single specimen.

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TABLE 4. Results of Ballistic Bend Tests.
(Specimens 1/2" x 1-1/2"x4"; Izod Notch on all 4 sides)

Plate, and Specimen Axis	Number of shots.	Rupture speed, F.P.S.	Bullet Energy, ft.lbs.	Nature of Fracture.
A, X	8	796 ~~4~~ 6	1398	Ductile; many deep pores.
B, X	4	678 ~~7~~ 5	1016	" " " "
C, X	4	650 ~~7~~ 5	932	" " " "
D, X	4	712 ~~7~~ 5	1118	" " " "
E, X	4	350 ~~7~~ 10	271	Mainly intercrystalline; many cracks; rough.
Sl, X	13	672 ~~7~~ 5	993	Ductile; large lip-structures and deep pores.
A, Y	4	685 ~~4~~ 10	1036	Ductile; with lip-structures.
B, Y	4	612 ~~7~~ 5	826	" " " "
C, Y	4	602 ~~7~~ 5	800	" " " "
D, Y	3	652 ~~7~~ 8	938	" " " "
E, Y	4	318 ~~7~~ 8	223	Mainly intercrystalline; many cracks; rough.
Sl, Y	4	628 ~~4~~ 16	867	Ductile; with lip-structures.
H, X	7	668 ~~4~~ 9	985	Ductile.
K, X	5	505 ~~7~~ 20	562	Fine mixture of ductile and cleavage fracture. In some places, bands of ductile fracture .020"- .040" wide.
M, X	5	365 ~~4~~ 15	294	Cleavage.
N, Y	4	374 ~~4~~ 10	309	Cleavage.
S2, Y	4	641 ~~4~~ 10	904	Ductile; with lip-structures.

TABLE 5. Comparison, of Static and Ballistic Bend Tests.
 Static Specimens 3/8" x 3/8"
 Ballistic Test Figures are for Specimens 1/2" x 1-1/2"
 S1 and S2 (the original treatment) are included for
 comparison.

Plate and Spec. Axis	Product	T.S. 10 ³ PSI (OX)	Static Work Dens. ft.lbs./ per.cu.in.	% of High- est Value.	Bullet Energy ft.lbs.	% of Highest Value.
<u>Tempered Martensites.</u>						
AX	Fine-grain	102	2748	(100)	1398	(100)
BX	" "	110	2308	84	1016	73
CX	" "	117	2037	74	932	67
DX	Coarse-grain	109	2241	82	1118	80
EX	" "	112	1952	71	271	19
S2X	(Temper Brittle) Fine-grain.	111	2341	85	-	-
AY	As above.	As	2402	(100)	1036	(100)
BY		above	1969	82	826	80
CY			1716	71	800	77
DY			2077	87	938	90
EY			1731	72	223	22
S2Y		111	1686	70	904	87
<u>Isothermal Products; Coarse-grain.</u>						
IX	400° product.	111	1819	77	985	99
KX	575° product.	112	2370	(100)	562	57
MX	625° product.	106	1972	83	294	30
SLX	Fine-grained.	110	2332	98	993	(100)
NY	625° product.	108	1482	-15	309	36
SLY	Fine-grain.	110	1741	(Base)	867	(100)

List of Useful Comparisons:

Grain-Size at same tensile strength:
 Temper Brittleness at same tensile strength:
 Hardness:
 Effect of re-heat treatment, see especially:
 Effect of isothermal transformation:
 Effect of specimen direction:

B, D
 D, E
 A, B, C
 S₂, B, D
 S₁, H, K, M, N
 AX, AY, etc;
 MX, NY.
 S₂X, S₂Y
 S₁X, S₁Y

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TABLE 6. Precracking Work and Postcracking Work,
Carnegie-Illinois Plate 52151; Treatment S1.
Energies in Foot Pounds.

Static Tests:

Specimen	Work,		E Total	Max. Load. Pounds	E_b/E
	E _a Precracking	E _b Postcracking			
S14-2	36	131	167	1548	
-4	40	143	183	1582	
-7	44	128	172	1604	
-11	42	132	174	1570	
Mean	44 ± 2	133 ± 2	174 ± 3	1576	76%
S14-1	44			1600	
-3	42			1584	
-5	45			1574	
-9	42			1582	
Mean	43 ± 1			1585	

Ballistic Tests: (Limit Determinations):

Specimen Condition	Bullet Energy:		Calculated Energy Expended at Notch:		T_b/T
	T _b Postcracking	T Total	Postcracking	Total.	
S14-6A, 6B, 8B, 10A, and 10B.	-	1003 ± 14	-	443	
S14-2A, 2B, 4A, 4B, 7A, 7B, 11A, 11B.		984 ± 15		435	
Mean		993 ± 15		439	81%
S14-1A, 1B, 3A, 3B, 5A, 5B, 9A, 9B.		807 ± 19		357	

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TABLE 7.

BRINELL HARDNESS NUMBERS FOR
CARNEGIE-ILLINOIS STEEL PLATE 52151

(All impressions centered in thickness direction.)

<u>Specimen No. (1-7/8" size)</u>	<u>B. H. N. (3000 Kg.)</u>
SLX1	233
SLX2	234
SLX3	236
SLY1	237
SLY2	237
SLY3	237
S2X1	229
S2X2	231
S2X3	231
S2Y1	232
S2Y2	233
S2Y3	231
H1X	232
H2X	232
H3X	230
K1X	229
K2X	226
K3X	225
M1X	215
M2X	215
M3X	215
N1Y	220
N2Y	221
N3Y	215

TABLE 8. Test of the Work Equation $E = Ad^3 + Bd^2$ on Bend Specimens of Carnegie-Illinois Plate 52151.

(E = rupture work, ft.lbs.; d = retained depth at notch in inches.)

Block Number and Specimen-Axis.	S1, OX	S2, OX	S1, OY	S2, OY	H, OX	K, OX	M, OX	N, OY
Treatment	As Received				400° Product	575° Product	625° Product	625° Product
A	1110	1157	756	858	858	1255	819	509
R	562	562	431	371	424	513	480	408
d = 3/8" Calc. Obs.	81 81	79 79	59 59	57 57	61 61	80 80	66 66	51 51
d = 1-1/8" Calc. Obs.	1267 1248	1269 1241	900 939	926 966	969 963	1331 1334	988 932	702 700
d = 1-7/8" Calc. Obs.	5016 5092	5088 5206	3523 3358	3729 3561	3848 3873	5392 5375	3852 4088	2637 2645

Work-Density Values in

ft.lb./cu.in.

Observed for 1-7/8"

specimen:

Calculated for

infinite specimen:

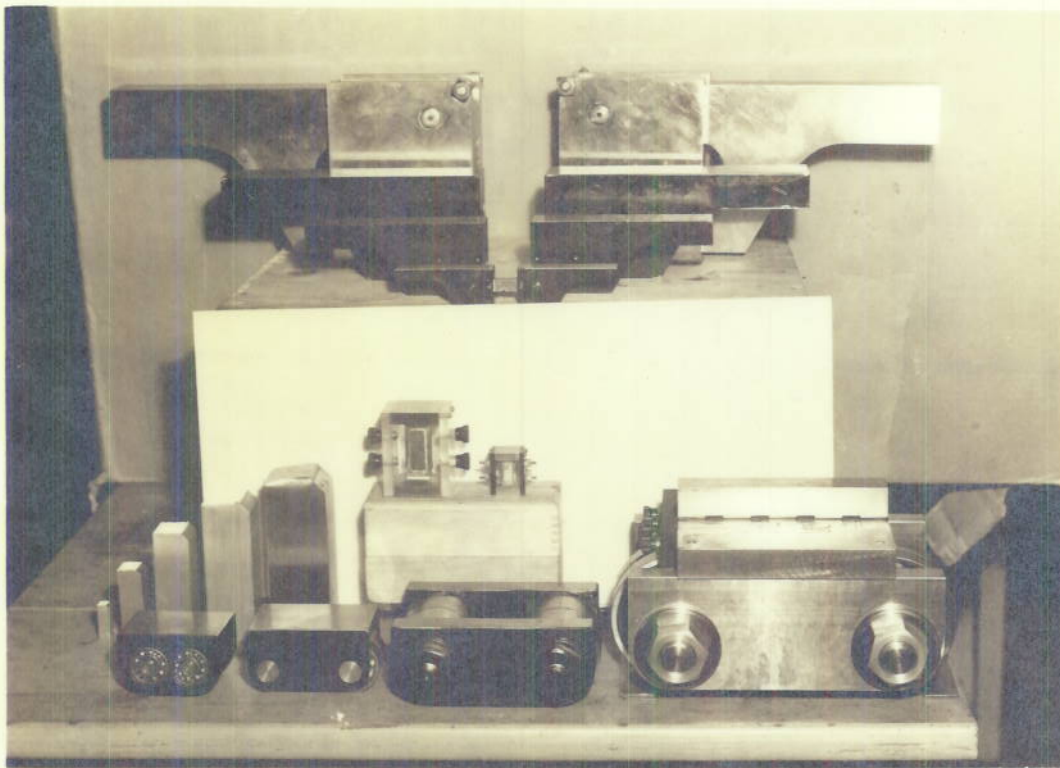
1207	1234	796	844	911	1274	955	641
888	926	605	686	728	1119	764	513

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TABLE 9. Values of the Constants in the Equation
for the Work per Unit Area $E/A = CA^n$.
(E in ft.lbs., A in sq.in. Only specimens with ductile
fracture are included.)

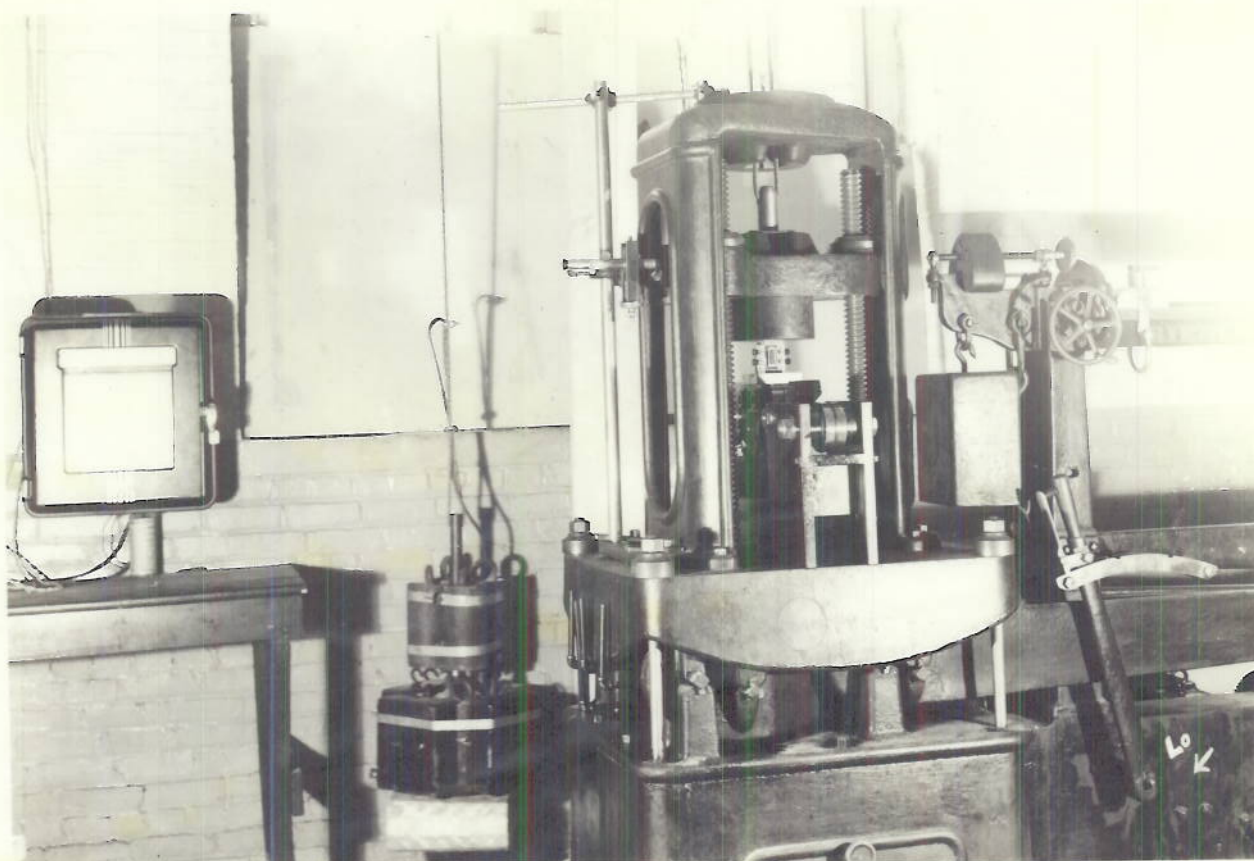
Treatment of Steel 52151.	Total work		Work Before Maximum Load		Work After Maximum Load	
	C	n	C	n	C	n
SLX	1300	.28	720	.33	578	.23
S2X	1320	.29	735	.32	570	.24
AX	1580	.30	930	.32	645	.26
BX	1345	.31	755	.31	584	.30
CX	1175	.30	632	.30	533	.28
DX	1340	.32	750	.34	604	.30
HX	990	.27	522	.24	470	.31
KX	1365	.30	766	.30	586	.28
SLY	916	.26	526	.28	390	.23
S2Y	942	.28	544	.34	398	.22
AY	1315	.27	812	.30	500	.24
BY	1165	.27	630	.27	430	.26
CY	937	.27	540	.29	380	.26
DY	1125	.27	663	.27	462	.27



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FIG. 1

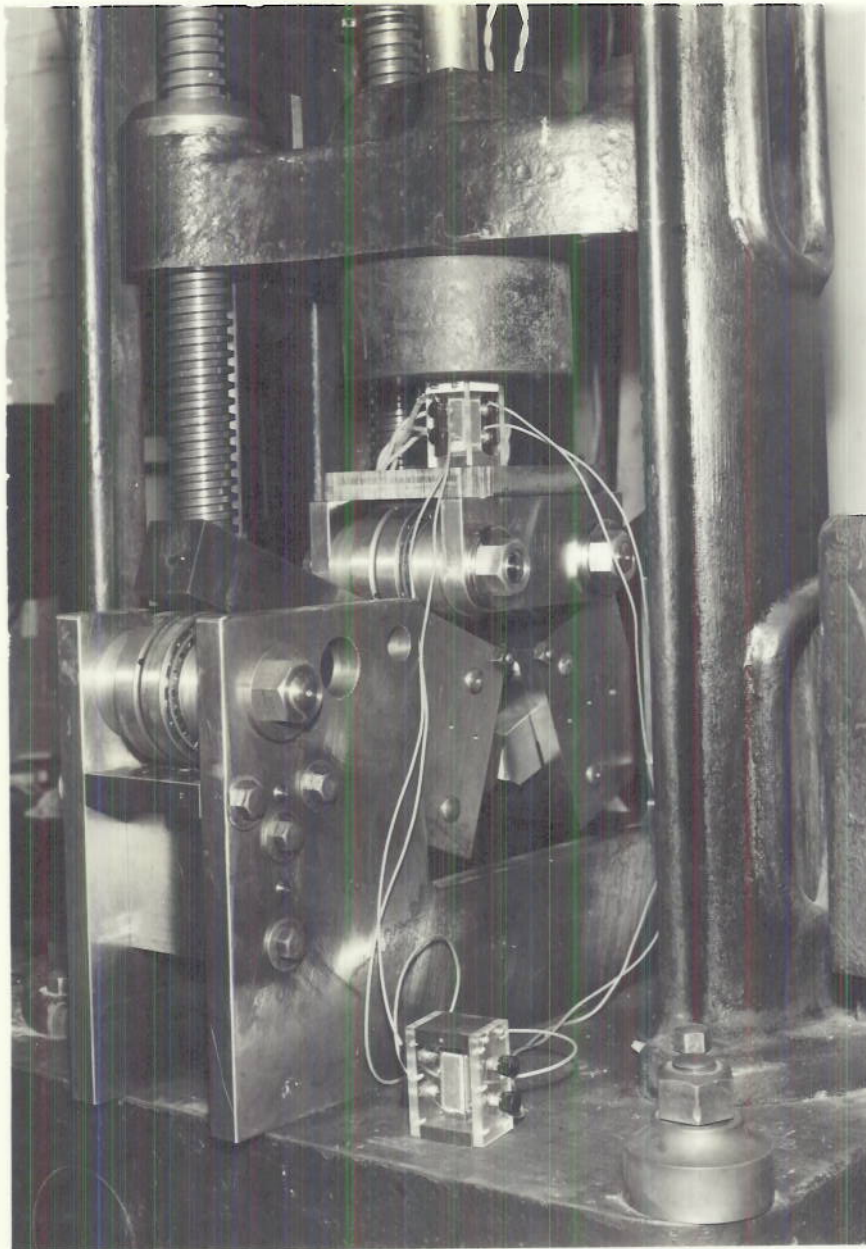


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FIG. 2

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DEC

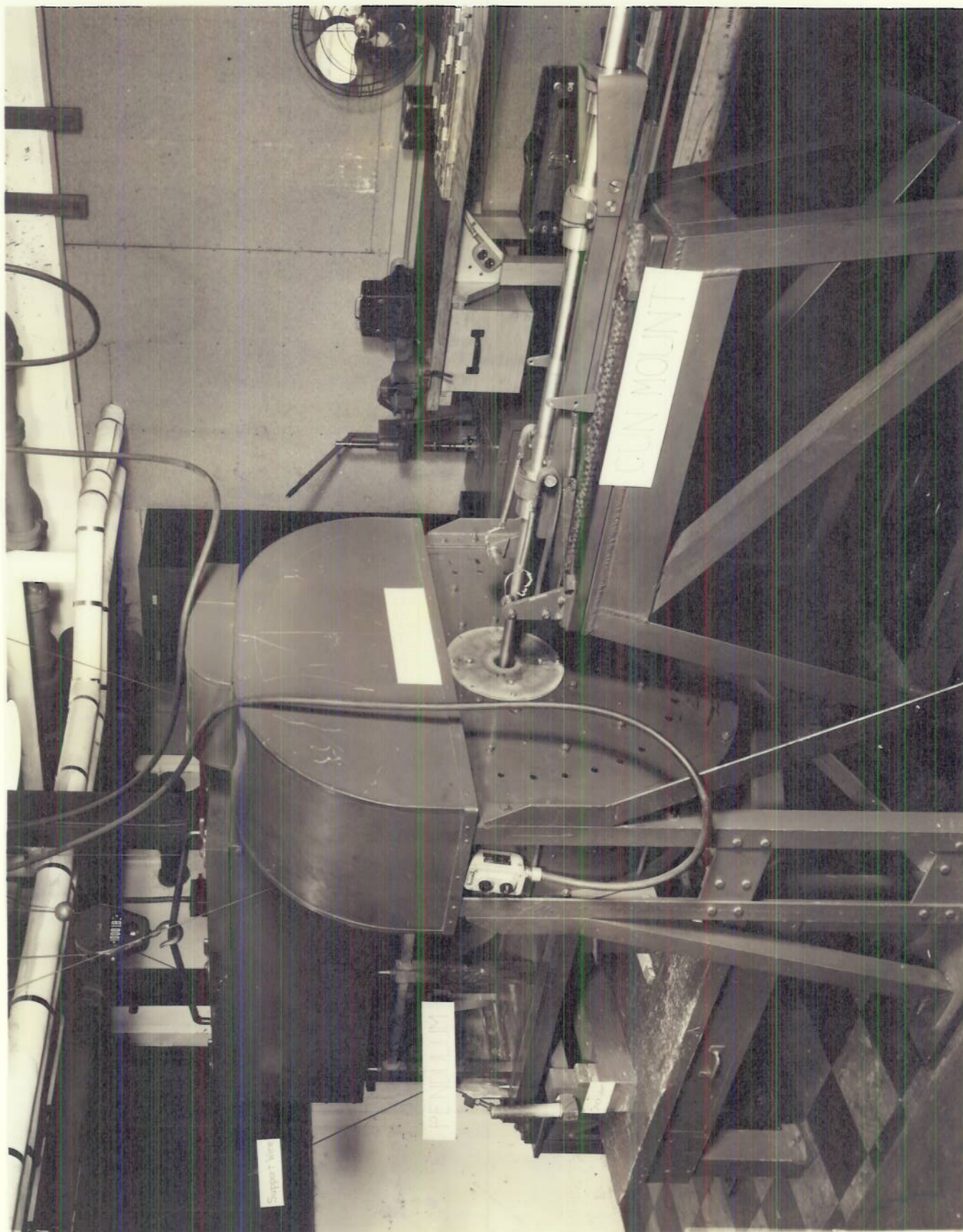


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FIG. 3



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FIG. 4

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FIG. 5

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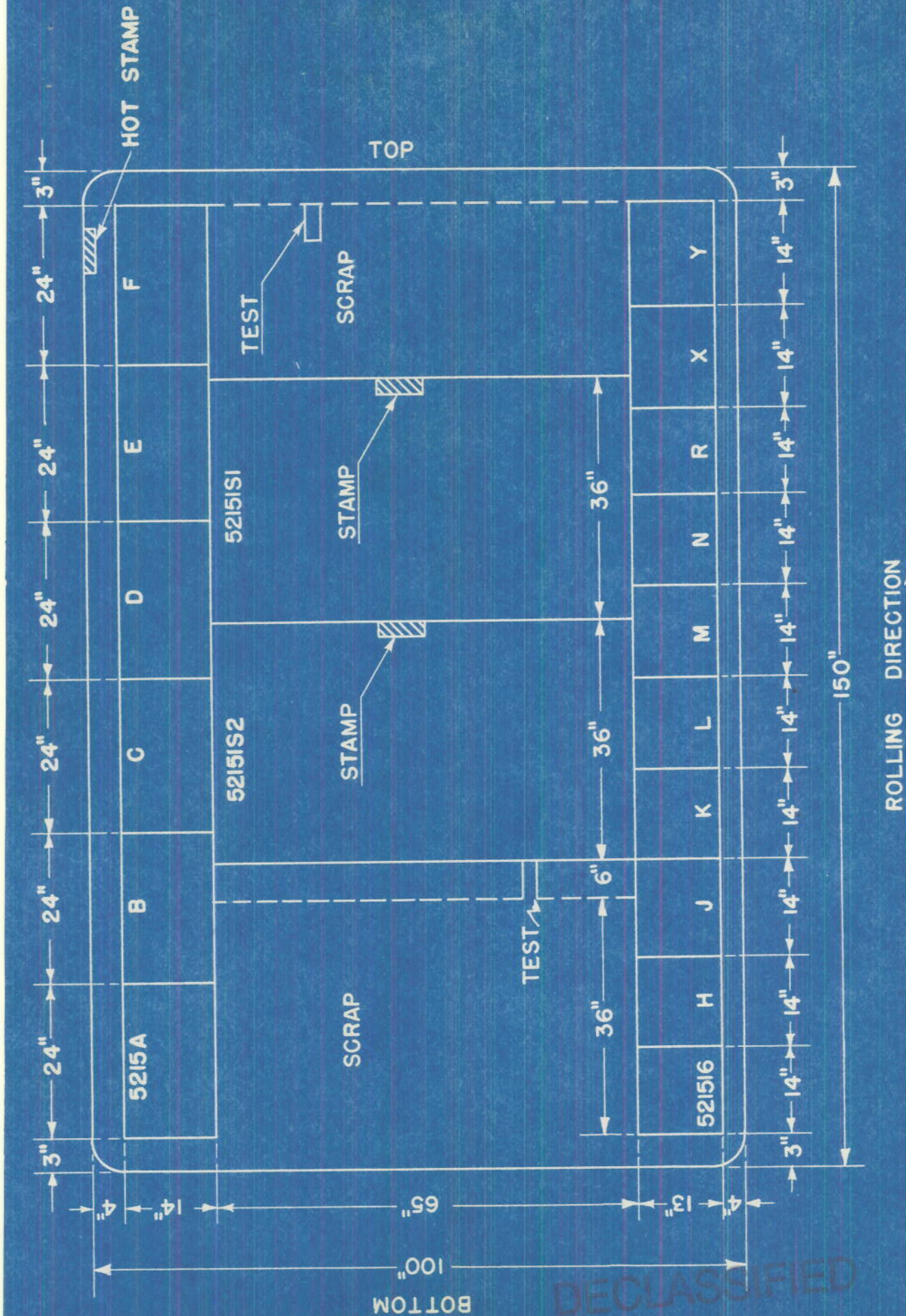


SPECIMENS AND PROJECTILES FOR BALLISTIC BEND TESTS ON STEELS.

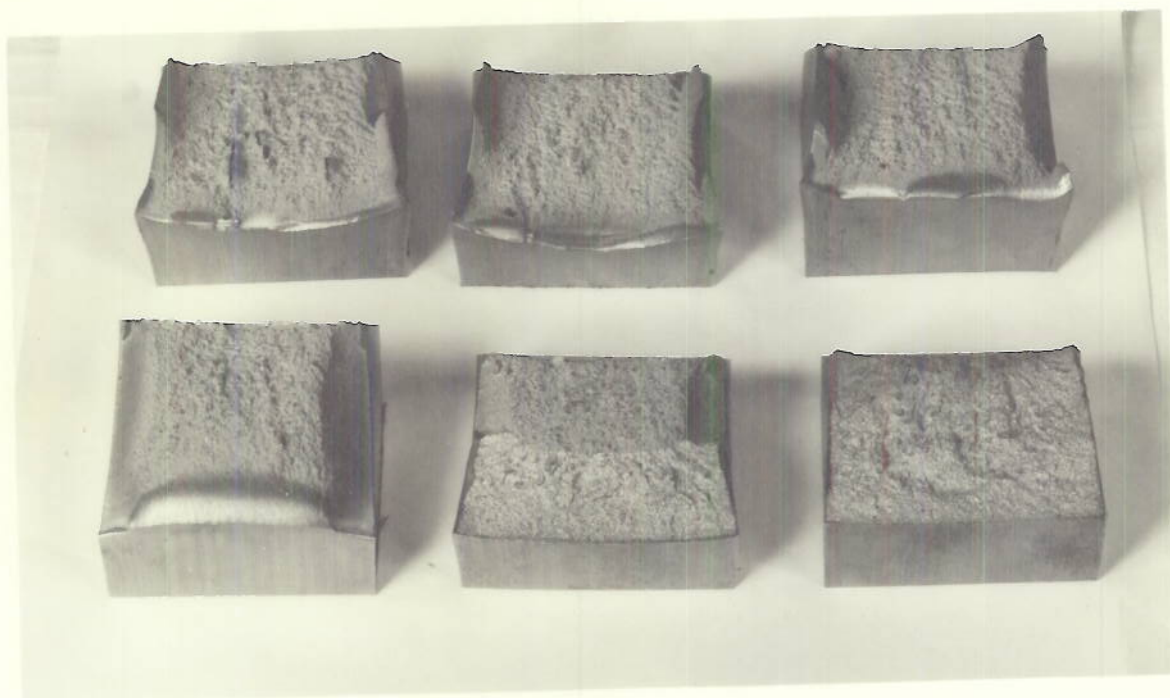
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FIG. 6



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FRACTURES OF STATIC BEND SPECIMENS, PLATE 52151.

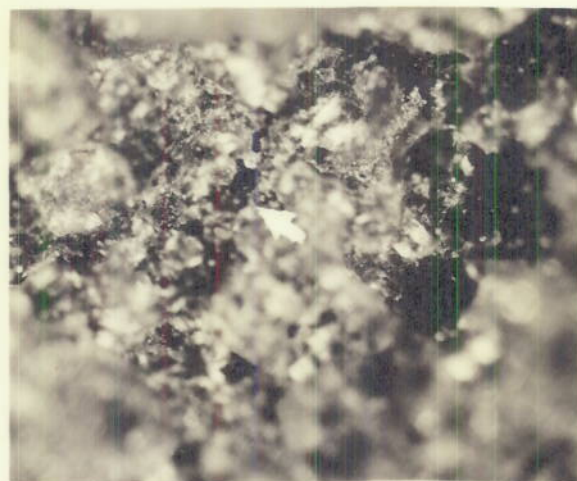
READING FROM UPPER LEFT, THEY ARE:

SIX, AS-RECEIVED MATERIAL; BX, FINE GRAINED, RETEMPERED;
HX, 400°C PRODUCT; KX, 575°C PRODUCT; MX AND NY, 626°C PRODUCT.

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FIG. 8



FRACTURES OF TEMPERED MARTENSITE IN A TEMPER BRITTLE CONDITION
PLATE 52151, TREATMENT E. MAGNIFICATION 50.

INTERGRANULAR FRACTURE
PERPENDICULAR TO FIBER

DUCTILE-FRACTURE AREA
PERPENDICULAR TO FIBER
IN THE SAME SPECIMEN

INTERGRANULAR FRACTURE
PARALLEL TO FIBER
(ARROW SHOWS AUSTENITE-GRAIN
PSEUDOMORPHS OF AVERAGE SIZE.)

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FIG. 9

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Traced from autographic sheets and
reduced to half scale.

Specimens from Plate 52151, sub-
plate M.

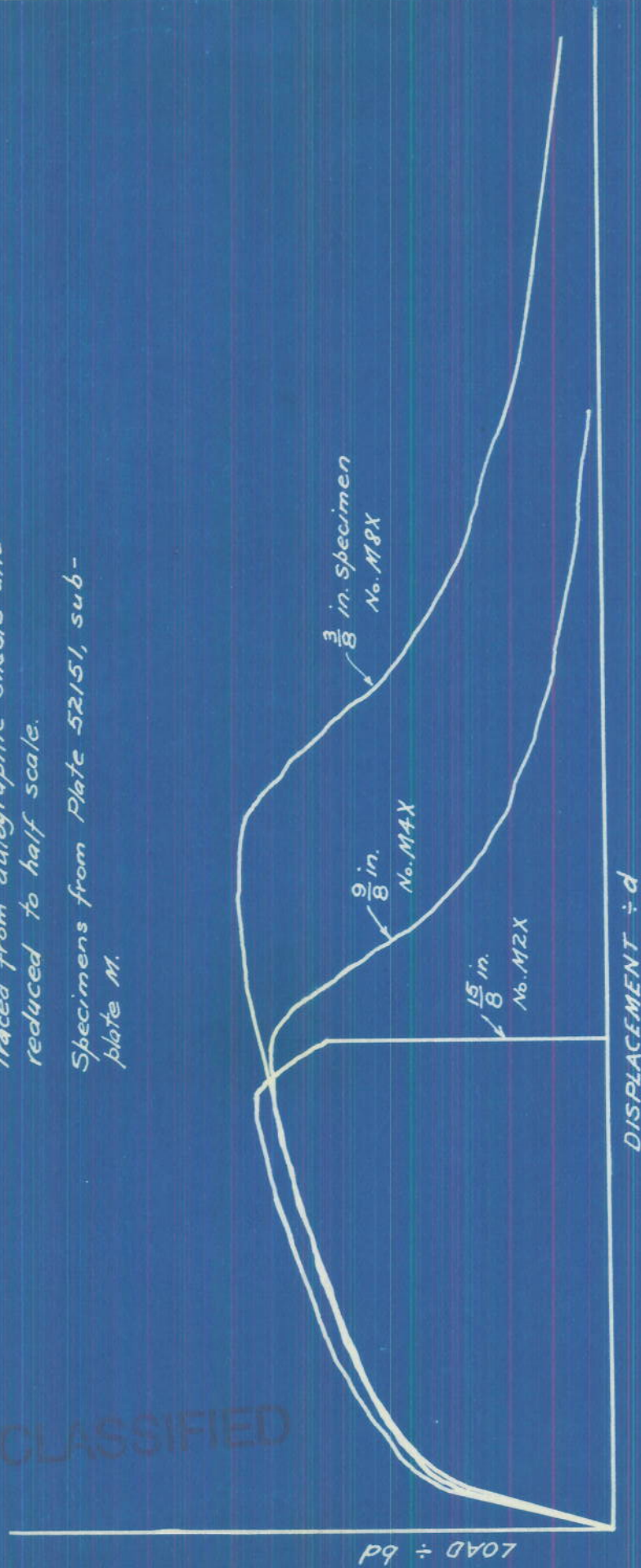
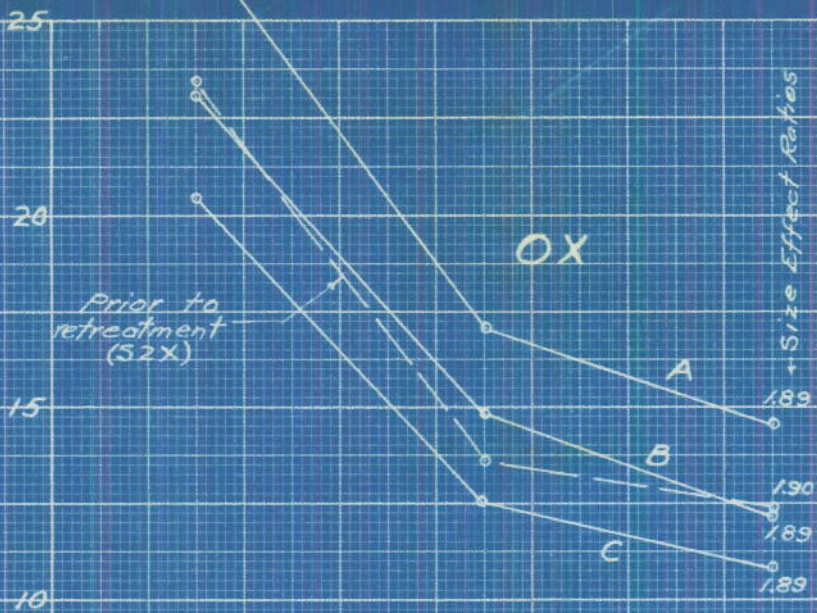


FIG. 10. AUTOGRAPHIC RECORDS. TYPICAL STATIC BEND TESTS.

0-2687

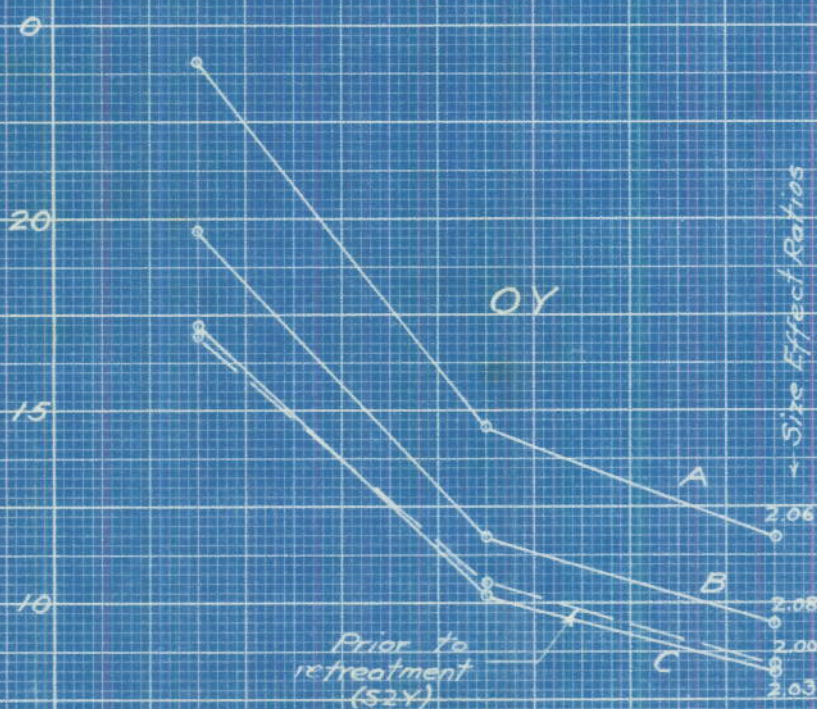
FIG. 10

(W) TOTAL WORK DENSITY - FT/LB PER CU IN $\times 10^{-2}$



Quenched and tempered to fine grained tempered martensite.

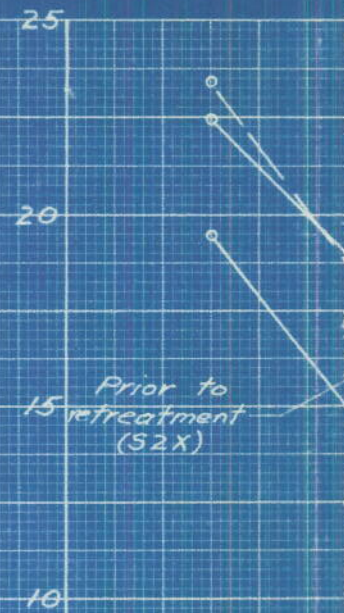
Plate	A	B	C	S2X	S2Y
BHN	214	237	252	230	232



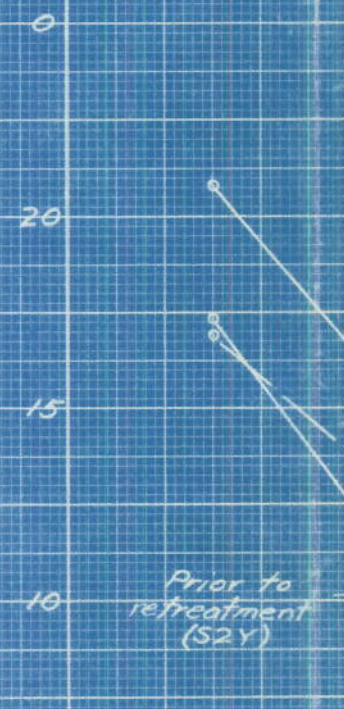
Spec. No. 1
Size, in. $\frac{3}{8}$

3
 $\frac{1}{8}$

5
 $\frac{1}{8}$

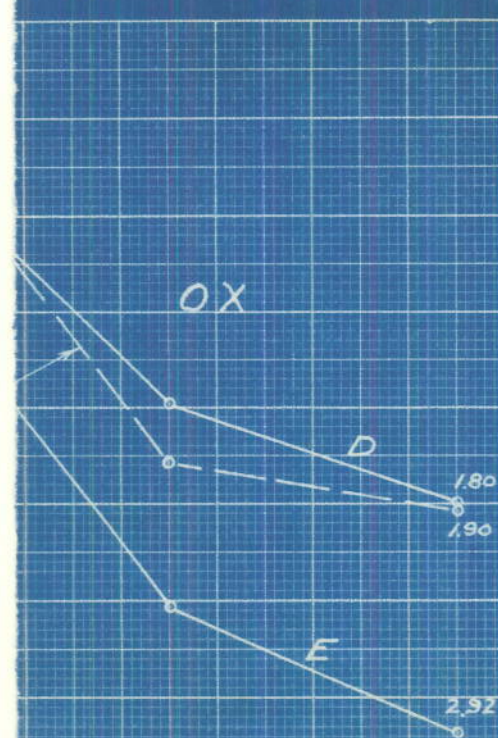


D. Quenched and grained per
E. Same and
BHN 233.

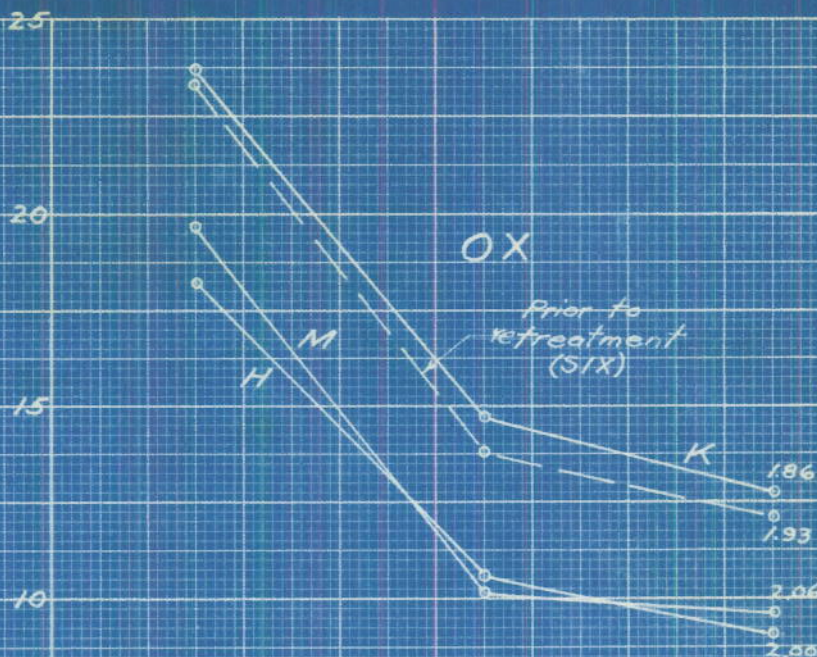


1
 $\frac{3}{8}$

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and tempered to fine
tempered martensite. BHN 226
rendered temper brittle.



H. Partially transformed to coarse
grained product 400°C.

K. Completely transformed to coarse
grained 575°C. product.

M, N. Same to 625°C. product.

Plate	H	K	M	N	SIX	SIX
BHN	231	227	215	219	234	237

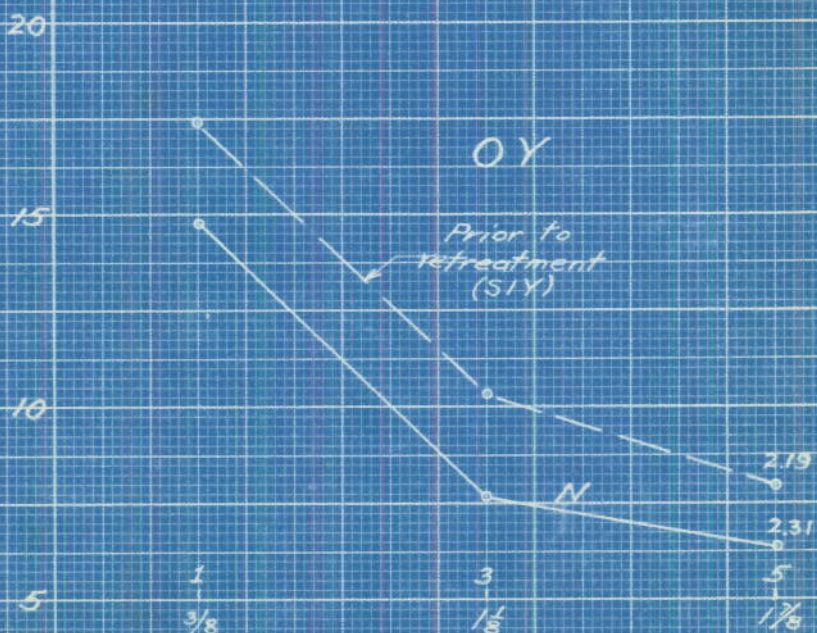
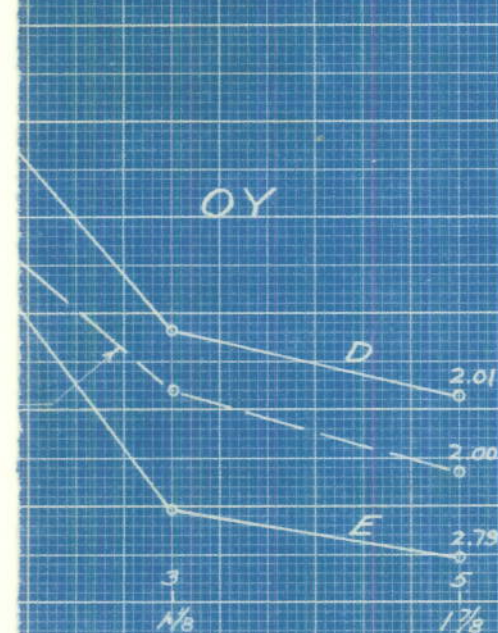


FIG. 11. REPORT 6-2687, NRL.
SIZE EFFECT IN W

(W.) WORK DENSITY UP TO MAXIMUM LOAD - Ft. lb./cu. in. $\times 10^{-2}$

20

15

10

5

0

OX

20

15

10

5

0

15

10

5

OY

15

10

5

Size No. 1

3

5

1

S₂X

A

B

C

S₂Y

A

B

C

S₂Y

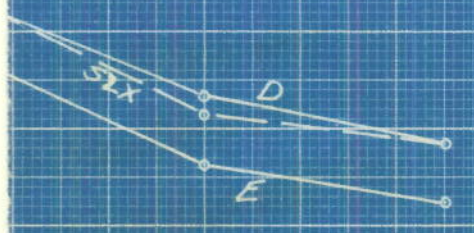
E

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L

FIG. 12. REPORT O-2687, NRL
SIZE EFFECT IN WL.

OX



20

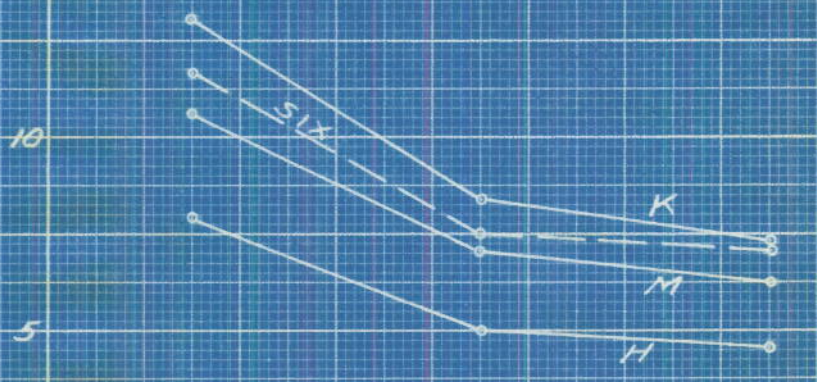
15

10

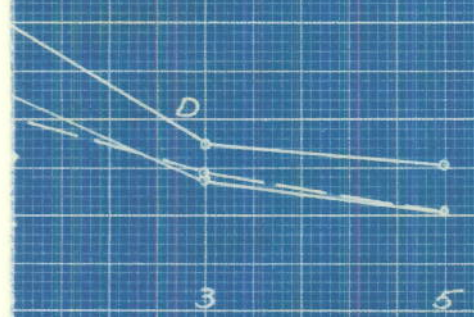
5

0

OX



OY

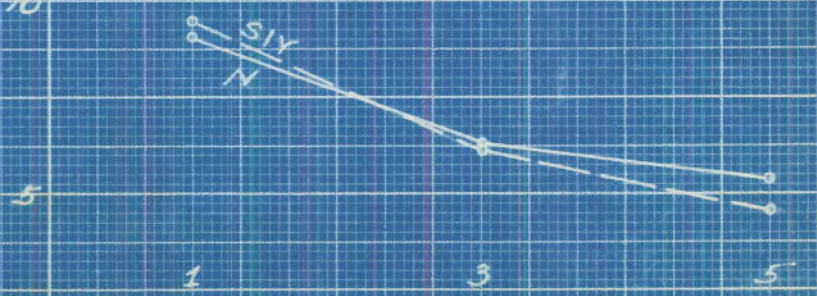


15

10

5

OY



(W₂) WORK DENSITY AFTER MAXIMUM LOAD - Ft./lb./cu.in. x 10⁻²

20

15

10

5

0

Size No. 1

3

5

OX

S2X

A

B

C

15

10

5

0

1

15

10

5

OY

C

S2Y

B

A

15

10

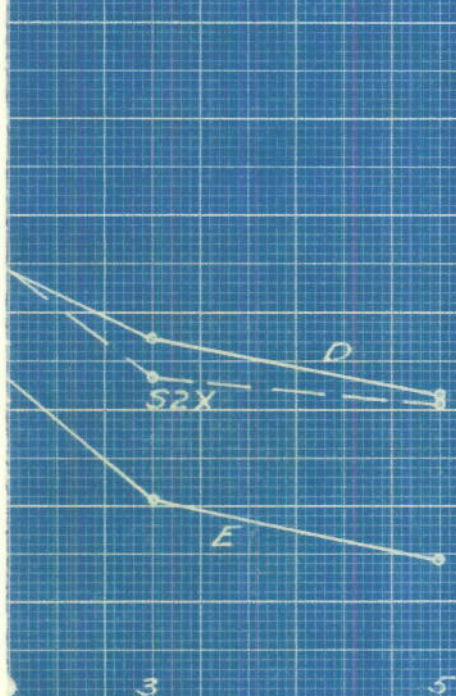
5

F

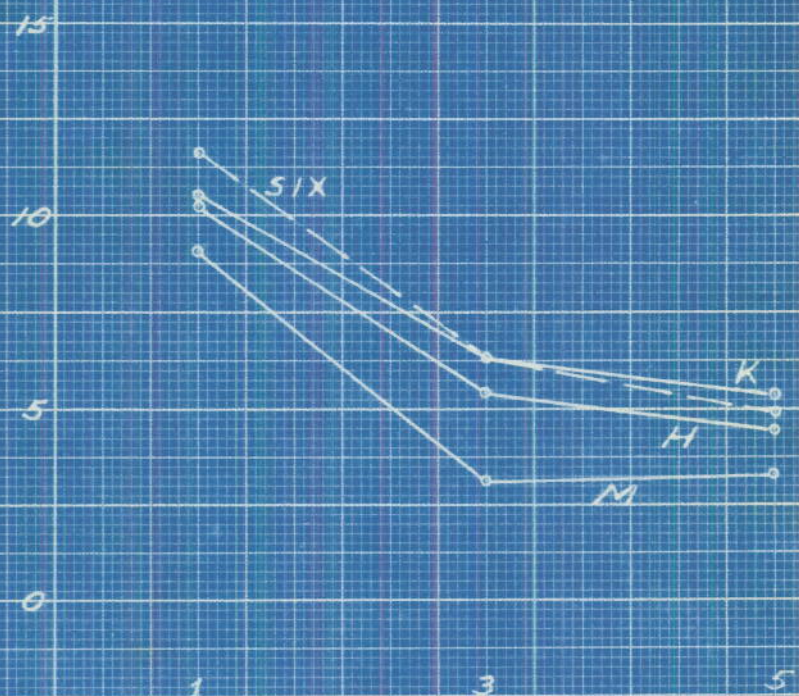
AL

FIG. 13. REPORT O-268T, NRL
SIZE EFFECT IN W_2

OX



OX



OY



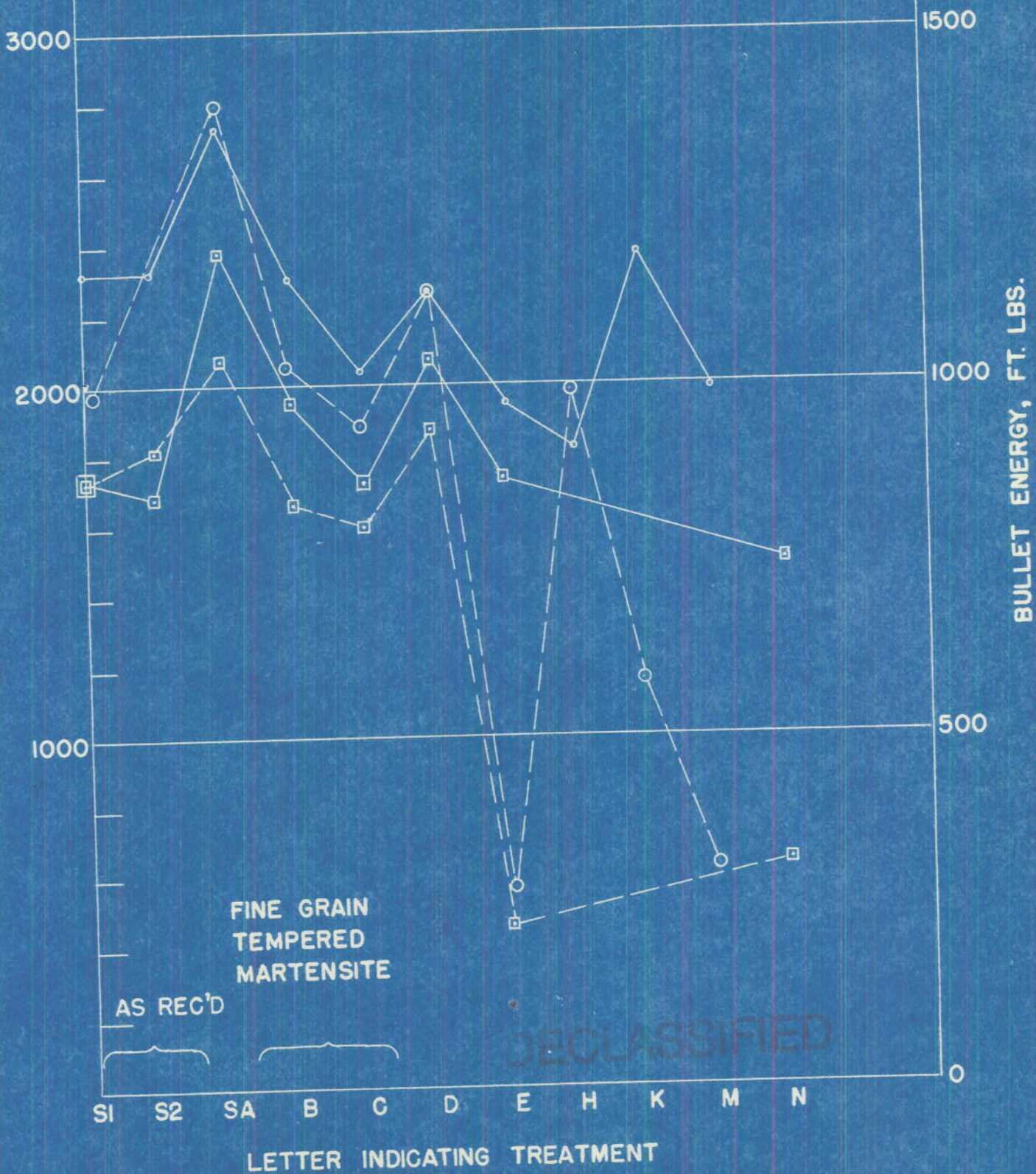
OY

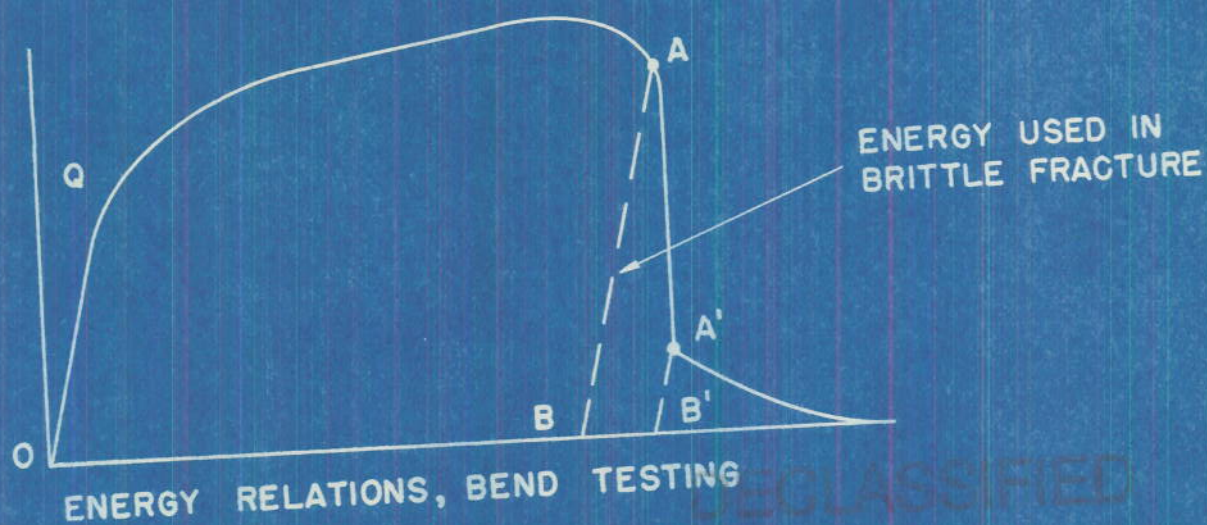
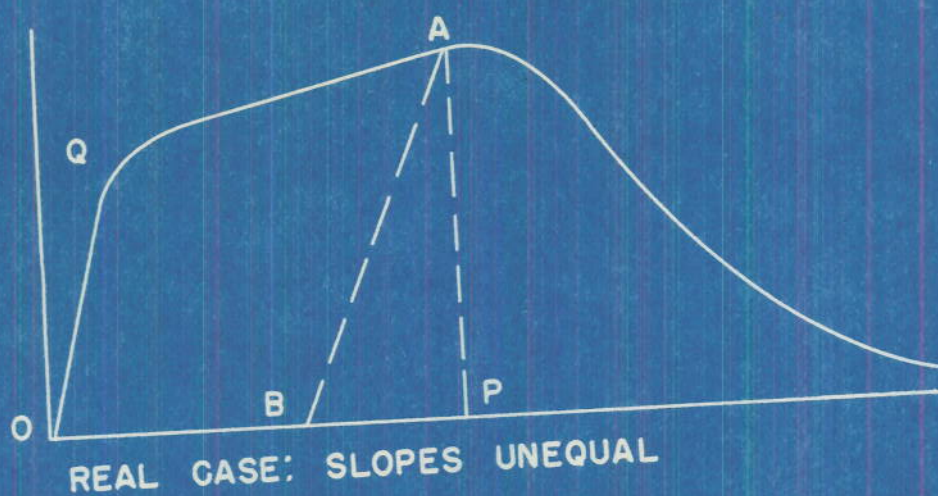
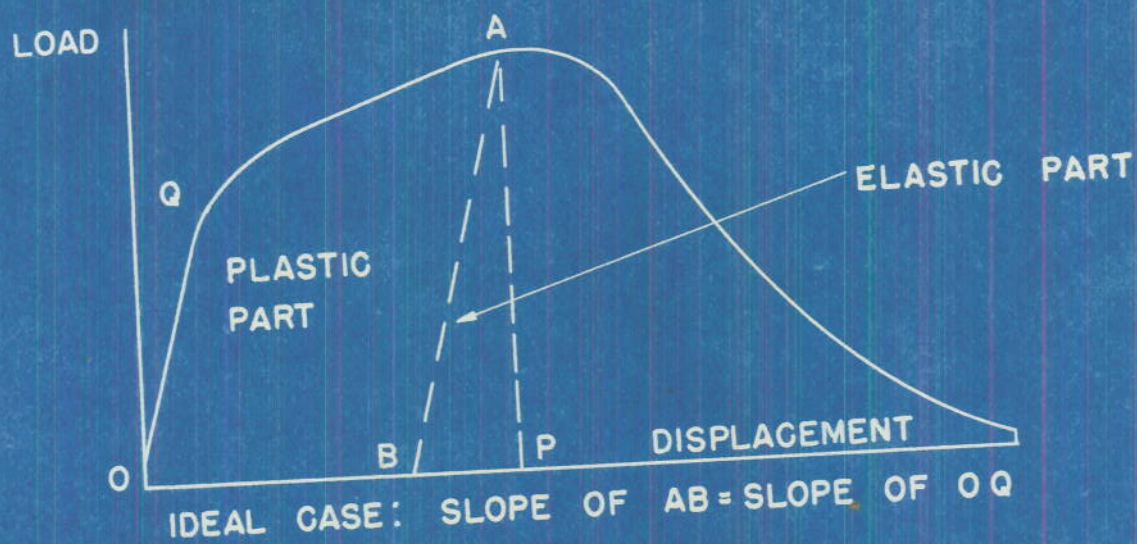


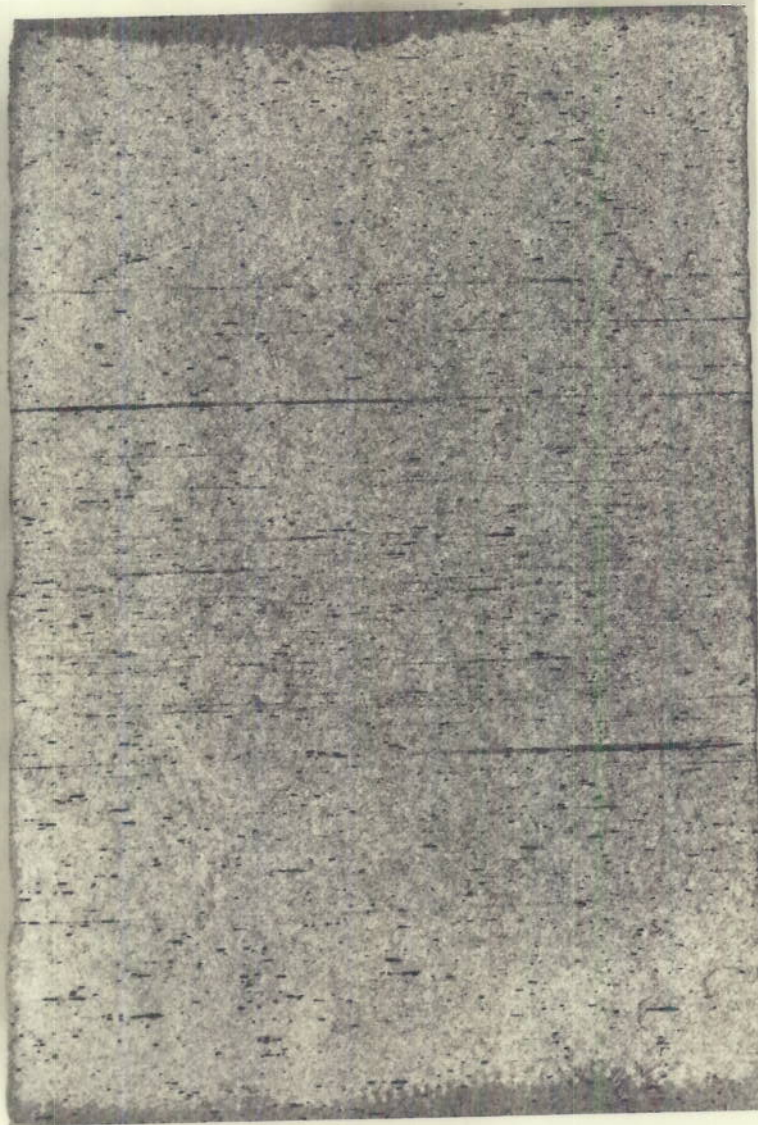
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COMPARISON OF STATIC AND BALLISTIC TEST, PLATE 52151

READ LEFT SCALE: — STATIC TEST, OX
 " RIGHT " : O — BALLISTIC TEST, OX
 READ LEFT SCALE: □ — STATIC TEST, OY
 " RIGHT " : □ — BALLISTIC TEST, OY



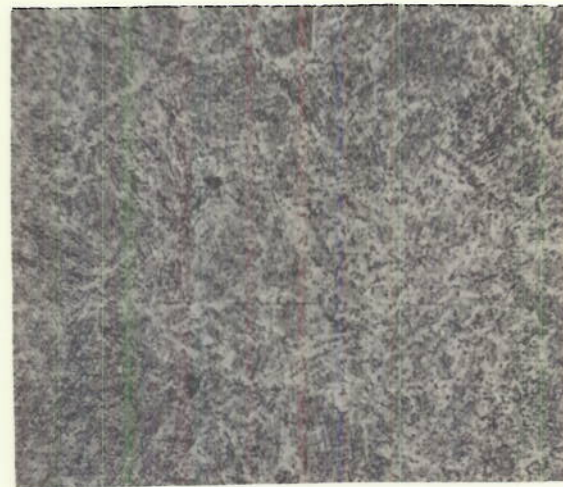
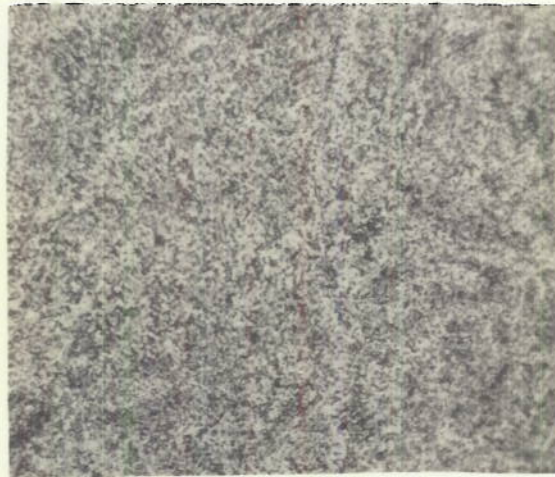
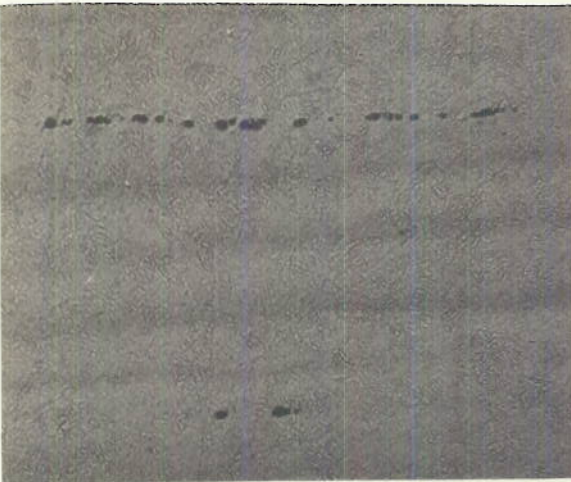




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FIG. 16

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100 X

1000 X

MICROPHOTOGRAPHS, PLATE 52151.

TOP S: FINE GRAINED TEMPERED MARTENSITE, AS RECEIVED.
 BOTTOM B: FINE GRAINED TEMPERED MARTENSITE.

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FIGURE 17A



100 X

1000X

100 X

1000 X

MICROPHOTOGRAPHS, PLATE 52151.

TOP H: 400°C PRODUCT.

MIDDLE J: 575°C PRODUCT.

BOTTOM N: 625°C PRODUCT.

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FIGURE 17B

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