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# FACTORS INFLUENCING THE STRENGTH DIFFERENTIAL OF HIGH STRENGTH STEELS

RICHARD CHAIT  
MATERIALS TESTING DIVISION

November 1971

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Technical Report by

*RICHARD CHAIT*

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Room temperature tensile and compressive true stress-true strain curves of various high strength steels (quenched and tempered 4340 steel, 410 martensitic stainless steel and H-11 steel; and aged 300-grade 18 Ni maraging steel) were analyzed to determine the effect of the various microstructures on what has been termed the strength differential (SD), i.e., the strength level difference between the tensile and compressive flow curves. Care was taken to insure that the compressive deformation was homogeneous. Regardless of the amount of plastic deformation, the quenched and tempered steels exhibited a higher flow stress in homogeneous compressive deformation than in tensile deformation. The extent of the SD was dependent on tempering temperature. This observation is consistent with what others have observed regarding yield strength behavior of quenched and quenched-and-tempered steels. Despite the low carbon content, aged maraging steel also showed a greater resistance to homogeneous compressive deformation. Metallographic examination of the maraging steel revealed the banding that is indicative of segregation. However, homogenization had little effect on the SD despite a change in austenite grain size, reverted austenite content, and the austenite-to-martensite transformational strains shown by Goldberg to be present in segregated material.

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## INTRODUCTION

To the designer, as well as to those attempting to characterize and understand mechanical working operations,<sup>1,2</sup> the uniaxial flow curve or true stress-true strain curve ( $\sigma$ - $e$ ) is of importance. While it is certainly desirable to be able to define the true or fundamental  $\sigma$ - $e$  curve, the fact remains that the appearance of the  $\sigma$ - $e$  curve may depend on the type of test that has been employed. With both tension and compression tests there are features which preclude uniform deformation through to fracture. Correcting for the influence of nonuniform deformation on the apparent flow stress does not insure agreement between the compressive flow curve,  $(\sigma-e)_c$ , and the tensile flow curve,  $(\sigma-e)_t$ . The observed strength level difference between the  $(\sigma-e)_c$  and  $(\sigma-e)_t$  curves has been termed the strength differential,<sup>3</sup> henceforth called SD.

Recent observations of the SD of high strength steels have been confined largely to the region of initial yielding.<sup>4</sup> Therefore, it is the objective of the present study to examine the SD over a wide range of plastic strain, and to indicate the influence of various microstructures on the SD. For this purpose, tensile and compressive flow curves of high strength steels with yield strengths of up to 300 ksi were obtained with techniques that minimized the influence of friction and tensile instability.<sup>5</sup>

## MATERIALS AND TEST PROCEDURE

Room temperature tension and compression tests were conducted on the following high strength steels: quenched-and-tempered 4340, H-11, and 410 martensitic stainless steel; and aged 300-grade 18 Ni maraging steel. The 4340 steel was heat treated to four different strength levels by tempering at 400 F, 600 F, 800 F, and 1000 F. The chemical compositions are given in Table I and heat treatments and resulting hardnesses in Table II.

### Compression Tests

The compression tests were conducted in a Baldwin Universal Testing Machine, using the apparatus schematically shown in Figure 1 to insure concentric loading. With the aid of aligning collars and centering washers, the amount of eccentricity was negligible as verified by the small difference among readings of 3 SR-4 strain gages placed 120° apart on an elastically loaded specimen.

To obtain the  $(\sigma-e)_c$  curve, cylindrical compression specimens, having a diameter of 0.333 inch and diameter-to-height ratio ( $d/h$ ) of 0.61, were used. Homogeneous deformation of the specimens was accomplished by coupling the renewal of the 0.002-inch-thick Teflon film lubricant with a remachining technique similar to that used by Taylor and Quinney.<sup>6</sup> The remachining operation closely restored the original  $d/h$  ratio. After each relubrication, the flow stress was obtained during the initial stage of plastic deformation (usually 2 to 3% reduction of height) to insure that neither peripheral breakdown of the Teflon film nor reloading effects would influence the compressive flow stress. Guided by a dial indicator, the strain rate was manually maintained as closely as possible to the  $3 \cdot 10^{-4}$  sec<sup>-1</sup> used in the tension testing portion of the program. Of major importance in determining the

Table I. MATERIALS AND THEIR COMPOSITIONS

	Steel				
	4340	Maraging	410	H-11 (Heat A)	H-11 (Heat B)
C	0.39	0.025	0.11	0.43	0.43
Mn	0.70	0.018	0.38	0.33	0.22
P	0.014	0.001	0.025	0.010	0.010
S	0.016	0.009	0.011	0.004	0.006
Si	0.29	0.01	0.49	0.94	0.99
Ni	1.77	18.52	0.43	—	—
Cr	0.92	—	12.49	5.04	4.88
Mo	0.25	5.08	0.03	1.20	1.37
V	—	—	—	0.52	0.52
Al	—	0.12	—	—	—
Co	—	8.90	—	—	—
Ti	—	0.62	—	—	—
Cu	—	—	0.11	—	—

Table II. HEAT TREATMENT AND RESULTANT HARDNESS OF MATERIALS

Steel	Austenitizing Conditions	Quenching Medium	Tempering Procedure	Average R <sub>c</sub> Hardness
4340	½ hr at 1550°F	oil	2 hr at 400°F, air cool	51
			2 hr at 600°F, air cool	48
			2 hr at 800°F, air cool	44
			2 hr at 1000°F, air cool	38
H-11	2 hr at 1850°F	oil	1 hr at 1050°F, air cool	49 (Heat A)
			(triple temper in this manner)	51 (Heat B)
Maraging	1 hr at 1500°F	air	3 hr at 900°F, air cool	53
410	1 hr at 1800°F	oil	4 hr at 800°F, air cool	41

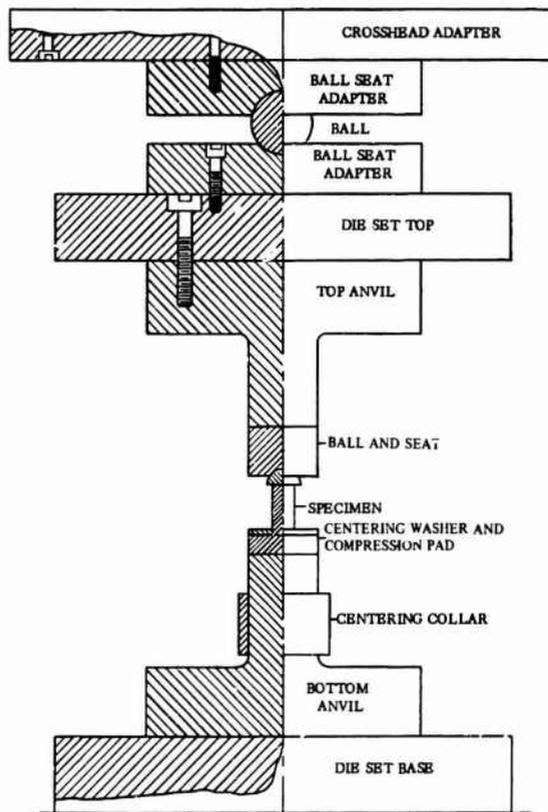


Figure 1. Compression testing apparatus

contribution of friction to the SD is the lubricant's effectiveness at the specimen-anvil interface. To determine the effectiveness of Teflon film lubrication, specimens with d/h ratios in the range 0.5 to 2.0 were given a small reduction in height, approximately 2%. The contribution of friction was ascertained from the dependence of the compressive flow stress on d/h.

For comparison purposes, nonhomogeneous compression was achieved by the unlubricated, uninterrupted compression of a cylindrical specimen (d/h = 0.61) of 4340 steel (600 F temper). An attempt was made to decrease the amount of nonhomogeneous deformation by testing specimens of 4340 steel (1000 F temper) of the following configurations: 1) broad-based or dumbbell-shaped specimens similar to that used in creep studies,<sup>7</sup> and 2) following Meyer and Nehl,<sup>8</sup> a layered cylindrical specimen consisting of three separate sections stacked together with MoS<sub>2</sub> powder between each section.

Both the  $(\sigma-e)_t$  and  $(\sigma-e)_c$  curves are plotted as true stress versus true plastic strain. The compressive strains were based on either the diameter of the largest cross section or, for those specimens that were remachined, on the specimen height. The true stresses for nonhomogeneously deformed compression specimens were based on the largest cross-sectional area.

### Tension Tests

Tension tests on the above materials were conducted in a 10,000-lb capacity Instron Testing Machine on specimens 0.200 inch in diameter. The instantaneous diameter was measured under load with a diameter gage similar to that described by Powell et al.<sup>9</sup> Finding and maintaining the location of the minimum neck diameter was done by manual movement of the diameter gage. A constant strain rate of  $3 \cdot 10^{-4} \text{ sec}^{-1}$  was achieved by following a diameter-versus-time master curve that corresponded to the desired strain rate.

The uniaxial flow stress after instability was obtained by dividing the true stress at the minimum section by a correction factor proposed by Beeuwkes,<sup>10</sup>

$$1 + (\pi/2) / [1 + (3\pi/2) / (e - e_n)] \quad (1)$$

where  $e_n$  is the strain at instability. In addition, tests were conducted where periodic elimination of the neck was achieved by means of split dies having the diameter of the minimum test section. The exact procedure consisted of removing the diameter gage and placing the dies and die holder around the necked area. This arrangement was used in conjunction with an hydraulic jack to reduce the material above and below the neck to the diameter of the minimum section. The specimen was secured in the Instron Machine with a light load during reduction. After the required reduction, the dies and die holder were removed and the diameter gage placed back in position. The test was then continued in the manner described above until the formation of another small neck.

### RESULTS

The  $(\sigma-e)_t$  and  $(\sigma-e)_c$  curves for 4340 steel are shown in Figure 2, and for 410 martensitic stainless steel, H-11 steel (Heat A), and aged maraging steel in Figure 3. Heat B of the H-11 steel is not shown in Figure 3 since curves similar

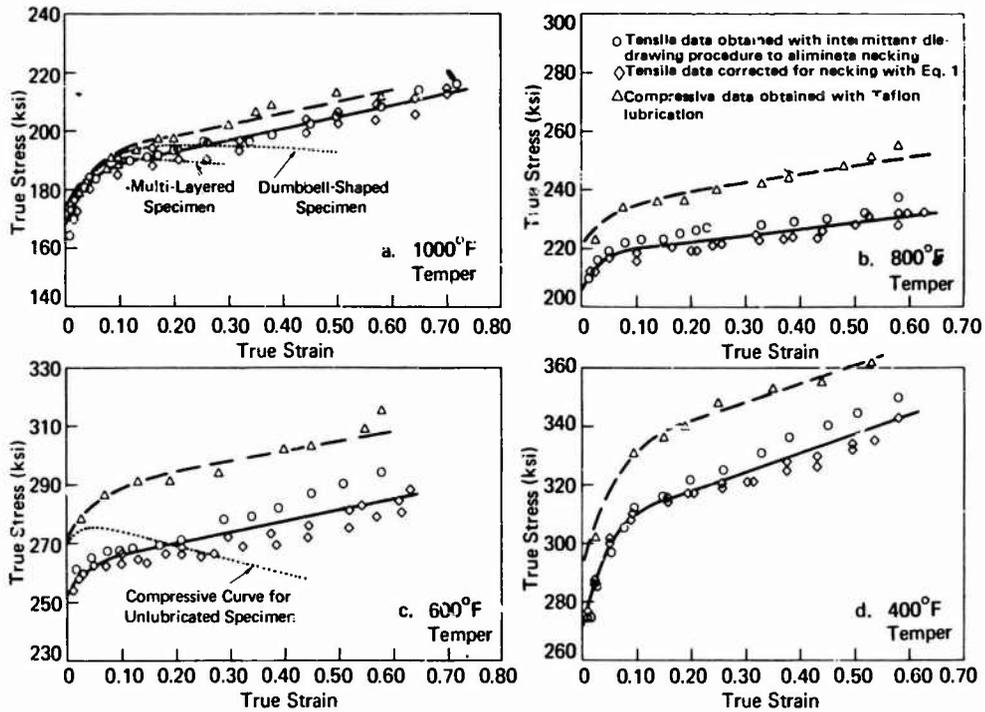


Figure 2. Room temperature tensile and compressive flow curves for 4340 steel quenched and then tempered at (a) 1000°F (b) 800°F (c) 600°F and (d) 400°F  
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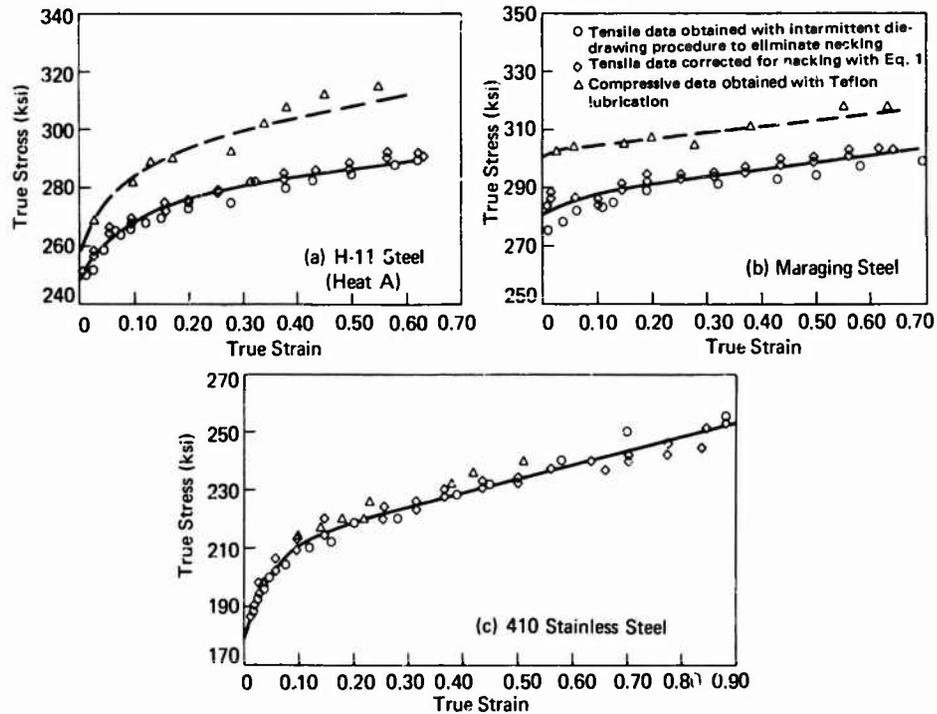


Figure 3. Room temperature tensile and compressive flow curves for (a) H-11 steel (heat A), (b) maraging steel and (c) 410 stainless steel  
19-066-1082/AMC-71

to Heat A were obtained. The data obtained from the ordinary tensile test, mathematically corrected for necking, agrees well with that obtained from tests where the neck was periodically eliminated by die-drawing. The  $(\sigma-e)_t$  curve represents a best fit through data points of these tests. To obtain SD, the difference between the  $(\sigma-e)_t$  and  $(\sigma-e)_c$  curves was noted at various locations along the curve. As shown in Figure 4, the SD of all materials remains at a relatively constant level despite large plastic deformation. Such behavior is particularly noteworthy as it bears on the applicability of some of the hypotheses that have been advanced to explain the SD. For purposes of comparison average SD values for all materials are shown in Table III.

With any compression test of this type, the question arises regarding the contribution of friction to the level of the  $(\sigma-e)_c$  curve. This contribution can be estimated with the analysis of Siebel<sup>11</sup> as detailed by Hill.<sup>12</sup> Here, the distribution of the compressive axial stress,  $\sigma$ , at a radius,  $r$ , is

$$\sigma = \sigma_0 \exp \left[ \frac{\mu(d-2r)}{h} \right] \quad (2)$$

where  $\sigma_0$  is the compressive flow stress under frictionless conditions and  $\mu$  is the coefficient of friction. Integrating Equation 2 to give the mean compressive flow stress,  $\sigma_c$ , gives

$$\sigma_c = \sigma_0 \left[ 1 + \frac{\mu d}{3h} \right] \quad (3)$$

This analysis implies that  $\mu$  is constant across the anvil-specimen interface. Van Rooyen and Backofen,<sup>13</sup> using instrumented pressure-sensitive pins in the compression platen to monitor the distribution of normal and shear stresses, found that Equation 2 described the stress distribution across the interface of an Al specimen lubricated with soap and reduced 2%. Pearsall and Backofen,<sup>14</sup> using the same technique, also noted essentially constant  $\mu$  across the interface of an Al specimen lubricated with Teflon film and reduced 20%.

In the present study, the technique of relubricating between small reductions can be expected to yield similar results. As shown in Figure 5, Equation 3 with  $\mu = 0.04$ <sup>15</sup> relates the variation in compressive flow stress to  $d/h$  for aged maraging steel. From this analysis, the frictional contribution to the measured flow stress is approximately 2 to 3 ksi for  $d/h = 0.61$ , the ratio used in obtaining the  $(\sigma-e)_c$  curves. Therefore, employing Teflon film together with frequent relubrication and remachining leads to essentially frictionless compressive deformation of high strength steels, a conclusion reached by Lee et al<sup>16</sup> for Al specimens relubricated with Teflon film but not remachined.

In contrast to the homogeneous deformation noted above, the appearance of the  $(\sigma-e)_c$  curve is changed considerably by the nonhomogeneous deformation (barrelling) that results from increased levels of friction. Barrelling becomes appreciable in an unlubricated, uninterrupted compression test, especially at large values of plastic strain. As a result of the nonhomogeneous deformation, the flow stresses decrease with increasing plastic strain (Figure 2c), thus supporting Polakowski's claim that the resistance to homogeneous deformation is greater than to nonhomogeneous deformation.

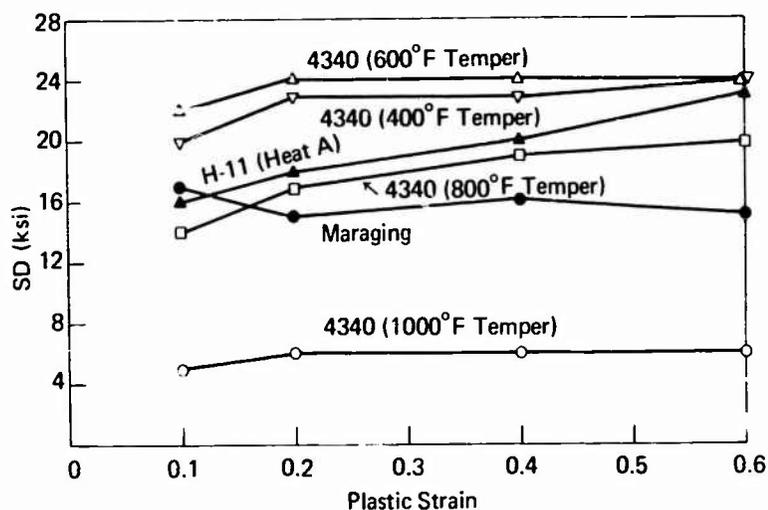


Figure 4. The effect of plastic strain on the strength differential

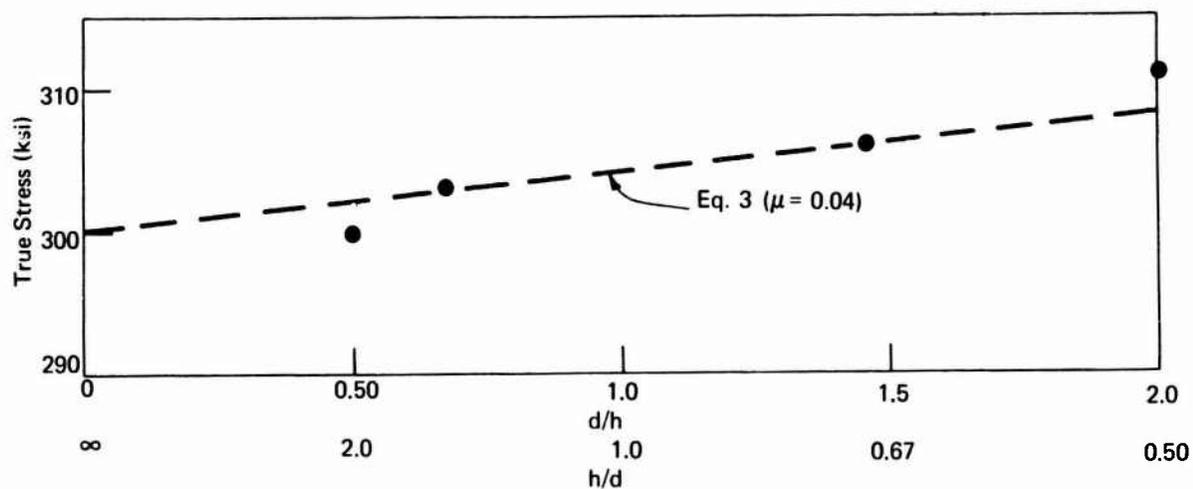


Figure 5. Effect of specimen dimension on the room temperature compressive flow stress of homogeneously deformed aged maraging steel (reduction in height approximately 2 percent)

Table III. COMPARISON OF THE STRENGTH DIFFERENTIAL LEVELS

Steel	Average SD (ksi)
4340 400°F Temper	22
4340 600°F Temper	23
4340 800°F Temper	17
4340 1000°F Temper	5
H-11 Heat A	18
H-11 Heat B	26
Maraging	16
410	Negligible

A variety of compression test specimens and procedures have been employed by other investigators to counter frictional effects, and as discussed by Polakowski,<sup>2</sup> these techniques do not necessarily provide homogeneous deformation. The attempts to eliminate barrelling by employing the dumbbell-shaped specimen and the specimen consisting of stacked units were not successful. As evidenced by the flow curves (Figure 2a) and the final specimen geometries (Figure 6), both types of specimens exhibited nonhomogeneous deformation.

## DISCUSSION

It is apparent that the frictional contribution to the flow stress does not account for the observed SD. Therefore, only a difference in the inherent response to tensile and compressive deformation can be responsible. For some metals and alloys, the mechanism behind the SD has been determined. Compressive deformation of a series of annealed Mg-Al alloys ranging from 0% Al to 9.14% Al involved substantial twinning.<sup>17</sup> Since twinning was minimal under tensile deformation, SD was appreciable. Twinning during low temperature compressive deformation of an Fe-0.15% Ti alloy had a similar effect.<sup>4</sup> The SD of TD nickel has been attributed to the tensile stress concentration at the particle-matrix interface.<sup>18</sup> Graphite flakes affect the deformation of certain cast irons in a similar manner.<sup>19</sup> The SD can also be traced to the influence of the mode of deformation on phase instability. For example, tensile deformation was found to promote more of the austenite-to-martensite transformation in austenitic stainless steels than compressive deformation.<sup>20</sup>

The cause of the SD in quenched or quenched-and-tempered steels, however, is still open to question. Because the magnitude of SD is not greatly affected by large plastic strain, it is highly unlikely that microcracking, residual stress or retained austenite contribute to the SD of quenched-and-tempered steels. The present study also indicates that the small amount of reverted austenite present in the aged maraging steel microstructure does not affect the SD. The latter observation and some additional hypotheses are discussed below.

### Quenched and Tempered Steels

One of the most important factors influencing the SD of quenched and tempered steels is heat treatment. As shown in Figure 7, the average SD of quenched-and-tempered 4340 steel as determined along the flow curve increases with decreasing tempering temperature. The yield strength values of 4340 steel displayed a similar trend.<sup>4</sup> Probably due to a specimen size effect,<sup>21</sup> the extrapolated value of the SD of as-quenched 4340 steel shown in Figure 7 is about half the value obtained by Leslie and Sober<sup>4</sup> for this percent C. It has been shown that the SD of as-quenched martensite will increase with increasing C content.<sup>4</sup> Therefore, it is not surprising to find that 410 martensitic stainless steel, with 0.11% C, exhibits negligible SD after tempering at 800 F. H-11 steel, despite the high tempering temperature, exhibits resistance to thermal softening and therefore displays a substantial SD.

There appears little likelihood that these observations can be explained by a Bauschinger effect induced by the austenite-to-martensite transformation.<sup>22</sup>

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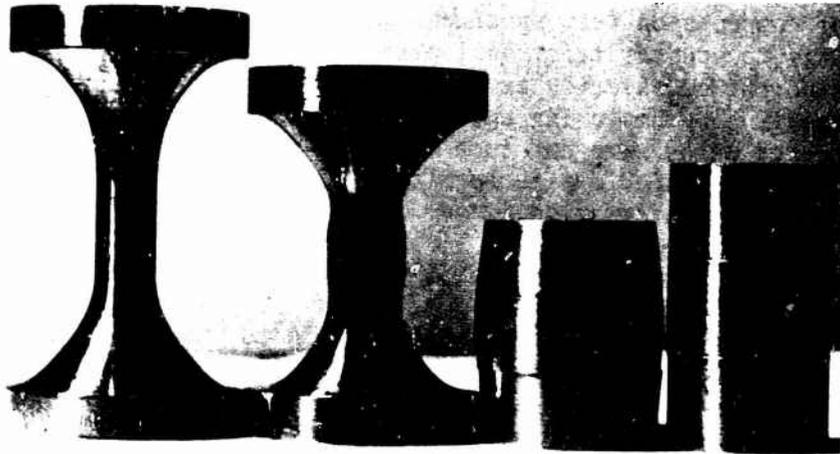


Figure 6. Nonhomogeneous compression of (a) dumbbell-shaped specimen and (b) multi-layered specimen

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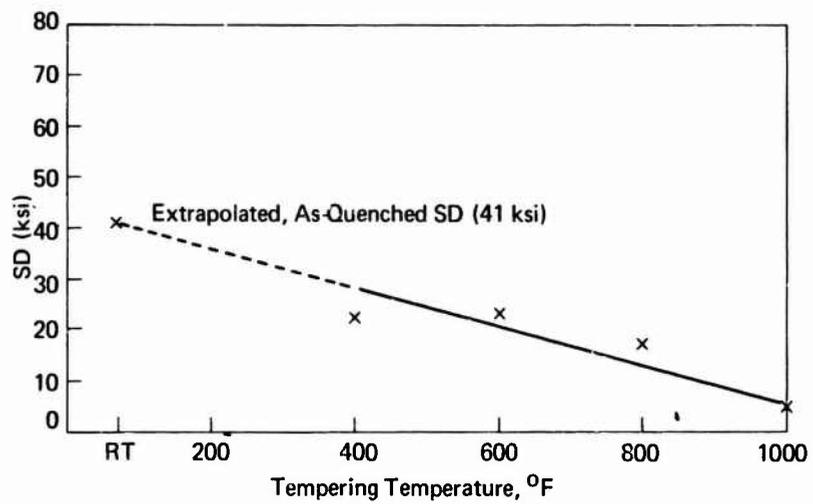


Figure 7. The effect of tempering temperature on the strength differential of quenched and tempered 4340 steel

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FACTORS INFLUENCING THE STRENGTH  
DIFFERENTIAL OF HIGH STRENGTH STEELS  
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More promising is the recent suggestion that the SD of as-quenched martensite could be interpreted in terms of a solute-dislocation interaction model where nonlinear elastic strains resulting from the distortion of the lattice around the interstitial atom lead to a greater interstitial dislocation energy in compression than in tension.<sup>3</sup> Precipitation of carbon from solution during the tempering process does not negate its application to structures of tempered martensite since the interaction may now involve carbon atmospheres adsorbed at dislocations.<sup>3</sup> Kalish and Cohen<sup>22</sup> have also used the nonlinear elastic treatment in proposing that lattice strains around coherently precipitated particles may lead to a greater resistance to dislocation motion under compressive deformation than under tensile deformation. It is possible that this mechanism may contribute to the SD of the H-11 steel since the formation of coherent precipitates of alloy carbides have been confirmed for 5 Cr-Mo-V steel tempered in the range 900 F to 1100 F.<sup>23</sup>

### Aged Maraging Steels

Despite its low carbon content, aged 300-grade maraging steel exhibits considerable SD even after extensive plastic deformation (Figure 3b). The ratio of the compressive to tensile flow stress along the flow curve is in the range 1.00 - 1.09 determined for aged 250-grade material in the region of initial yielding.<sup>24</sup> Therefore, the banded microstructure shown in Figure 8a does not appear to significantly affect SD. To pursue this point, the  $(\sigma-e)_c$  curve was determined for 300-grade material that was given an homogenizing treatment of 26 hours at 2190 F before solutionizing and aging. This homogenization treatment, used successfully by Goldberg<sup>25</sup> affects the microstructure of maraging steels by altering the reverted austenite content, austenite grain size, and the level of the austenite-to-martensite transformational strain. Each of these features and their effect on the SD are now discussed.

The appearance of the reverted austenite is promoted by chemical segregation that results from inadequate working and homogenization.<sup>26</sup> The amount of reverted austenite present in the banded microstructure of Figure 8a was approximately 8% as determined by the X-ray analysis described by Miller.<sup>27</sup> Therefore, it is not unexpected that adequate homogenization (26 hours, 2190 F) eliminated both the reverted austenite and the banding (Figure 8b) of the 300-grade material used in the present study. Despite the change in the microstructure, the  $(\sigma-e)_c$  curve of the homogenized material remained at substantially the same level as for the un-homogenized material. The tensile properties (yield and tensile strength) of homogenized and banded 250-grade material (aged) have also been reported to be about the same.<sup>28</sup> Therefore, it is concluded that banding and reverted austenite do not affect the SD of normally aged 18 Ni maraging steel.

It is possible to eliminate the reverted austenite and still leave bands in a 100% martensitic structure.<sup>25</sup> These bands, believed to result from segregation, are accompanied by anisotropic austenite-to-martensite transformational strains.<sup>25,29</sup> Extended homogenization eliminates both the bands and the transformational strains. However, it is not quantitatively known how the austenite-to-martensite transformational strains affect subsequent aging. Prolonged homogenization had no effect

on either tensile properties of aged 250-grade material<sup>28</sup> or as noted above, on the compressive properties of the aged 300-grade material of the present study. Therefore, it is concluded that transformational strains do not affect SD.

The same conclusion is reached regarding the effect of increasing austenite grain size. Prior to aging, the homogenized (26 hours at 2190 F) and solutionized (1 hour at 1500 F) 300-grade material of the present study had an ASTM austenite grain size of about 4, while unhomogenized material had an ASTM grain size of approximately 8. The lack of an effect on the compressive properties is attributed to a similarity in the martensite unit size<sup>30</sup> (Figure 9).

It is apparent that the effects of banding, grain size, and transformational strains do not contribute to the SD of normally heat-treated 18 Ni maraging steels. Therefore, to account for the SD, there must be an additional mechanism that is operative. Since the precipitation of some compounds has been observed to be coherent with the matrix,<sup>31</sup> it is possible that as with H-11 steel an elastic interaction<sup>22</sup> also contributes to the SD of aged maraging steels.

### CONCLUSIONS

The following conclusions can be drawn from this investigation, which has examined the strength differential (SD) of several high strength steels as determined from a comparison of uniaxial compressive and tensile flow curves:

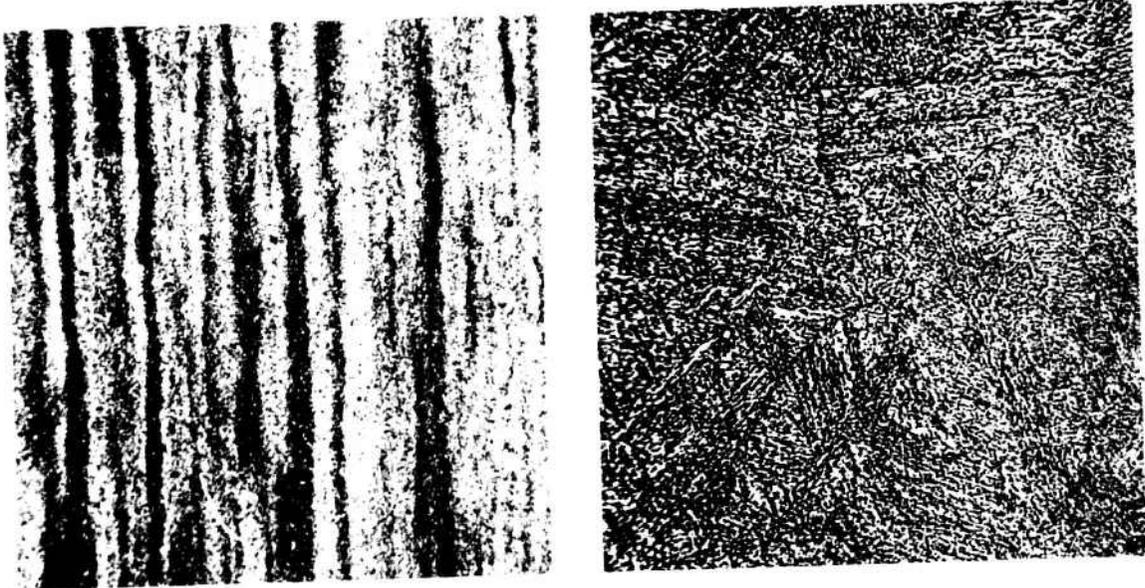
1. All steels, except 410 martensitic stainless steel, exhibited greater resistance to compressive deformation than to tensile deformation. When an SD was observed it remained essentially constant over a considerable range of plastic strain.

2. The contribution of friction to the compressive flow stress does not account for the presence of the SD.

3. The 18 Ni maraging steel, which depends on an aging reaction for strengthening, has an extremely low carbon content but nonetheless exhibits a SD that is comparable to some quenched and tempered steels. Homogenization, prior to heat treatment, has little effect on SD. This occurs despite the fact that (a) the microstructural features accompanying segregation (banding and reverted austenite) are eliminated, (b) austenite grain size is increased, and (c) austenite-to-martensite transformational strains are substantially reduced.

4. For the quenched and tempered steels, increasing the tempering temperature reduces the SD. Also, the existence of the SD at large values of plastic strain negates any contribution from microcracks, residual stresses, and retained austenite. A possible contribution to SD is the nonlinear elastic strain that stems from distortion of the lattice by interstitials resulting in a dislocation interaction that is greater in compression than in tension.<sup>3</sup>

5. Applicable to the SD of H-11 steel and aged maraging steel is the suggestion that coherently precipitated particles can also lead to nonlinear elastic strains.



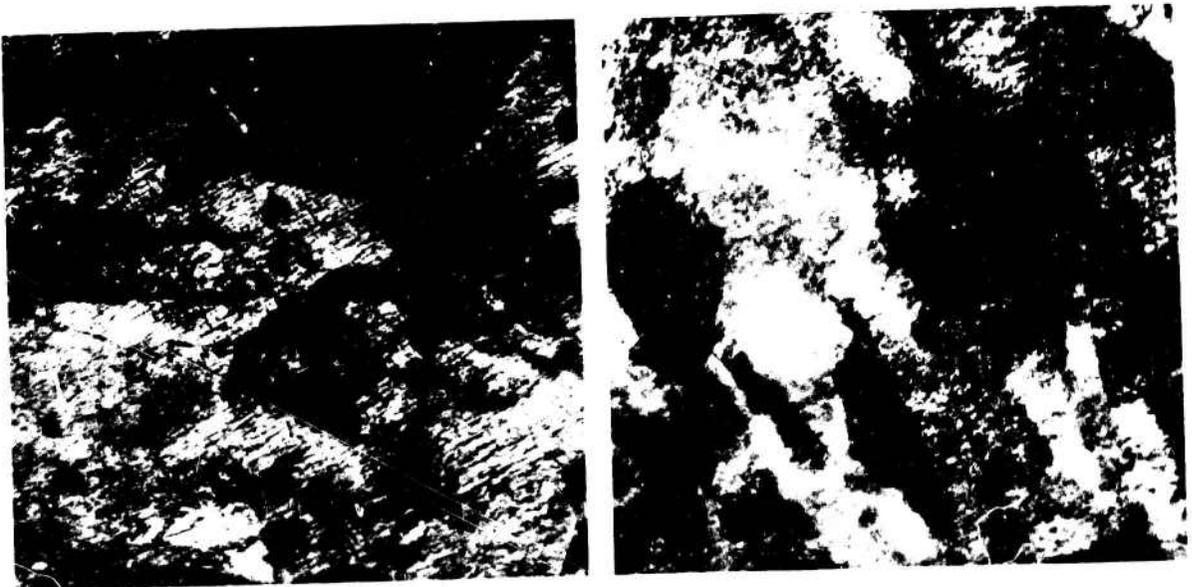
(a)

(b)

Figure 8. Microstructure of aged maraging steel (a) before and (b) after homogenization treatment. Mag. 50X

19-066-832/AMC-70

NOT REPRODUCIBLE



(a)

(b)

Figure 9. Transmission electron micrographs of aged maraging steel (a) before and (b) after homogenization treatment. Mag. 30,000X

6. Homogeneous compressive deformation that is essentially frictionless can be achieved by continually remachining and relubricating with Teflon film. Modifications of the cylindrical specimen did not prove successful in negating friction. As a result of the friction, the compressive flow curves fell below those representative of homogeneous deformation.

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