Prediction of Mechanical Properties of Multiphase Steels Containing Retained Austenite

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A theoretical model has been developed for predicting the flow behavior of dual-phase steels containing retained austenite. A good agreement between theoretical predictions and experimental values is obtained by optimizing the value of load-transfer factor and by using an additional term considered to account for the geometrically required dislocations around the hard particles.

1. Introduction

Many low-alloy steels have been processed in recent years to achieve various combinations of phases, such as ferrite, austenite, martensite, bainite, etc. Among these, ferrite-martensitic dual-phase and ferritic-martensitic-austenitic multiphase steels have been found to possess favorable strength elongation properties. Mathematical modeling of the microstructure-property correlation obviously considerably facilitates the alloy design and optimization of its properties. A number of such efforts[1-3] have been directed toward the modeling of the behavior of dual-phase/multiphase steels. Most of these models are based on continuum mechanics using various laws of mixtures for composite materials. The basic constitutive equations for such an approach have been written more succinctly by Kim[3] for two-phase materials, and these are as follows.

- The deforming system (the dual-phase material) must satisfy the following equations:

$$\sigma_c = \sigma_\alpha + qV_\beta\varepsilon_\alpha \qquad \text{(Eq 1)}$$

$$\varepsilon_c = (1 - V_\beta)\varepsilon_\alpha \qquad \text{(Eq 2)}$$

where the subscripts c, α, and β are the composite, matrix structure, and hard phase, respectively;
σ, ε, and V are stress, strain, and volume fraction of phases, respectively; and
q is the load-transfer factor.
These symbols will be used throughout the text.

- Plastic deformation of both phases.

$$\sigma_c = \sigma_\alpha + qV_\beta(\varepsilon_\alpha - \varepsilon_\beta) \qquad \text{(Eq 3)}$$

$$\sigma_c = \sigma_\beta - q(1 - V_\beta)(\varepsilon_\alpha - \varepsilon_\beta) \qquad \text{(Eq 4)}$$

$$\varepsilon_c = (1 - V_\beta)\varepsilon_\alpha + V_\beta\varepsilon_\beta \qquad \text{(Eq 5)}$$

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Equations 3 and 4 combine to yield the law of mixture for a two-phase material, but in the present form, they take into account, explicitly, the back stresses resulting from differential deformation of the two phases.

The choice of load-transfer factor, q, may make a difference between the success and failure of the above analysis; it is, therefore, appropriate to discuss its significance at this juncture.

2. Two-Phase Steels

Tomota et al.[4] and Kim[3] in their analyses used q values based on composite elastic considerations. Such considerations yield a very high q value and are nearer to the equal strain criterion for a composite. This may apply to materials where relative "flow ductility" of two phases is not very different. On the other extreme, a very low q value corresponds to equal stress criterion and may have justification where the hard phase has a very high strength but very low ductility as compared to the matrix.

In materials such as dual-phase steels, conditions between the above two extremes are likely to prevail. Although the relative strength of the two phases is quite different, even the hard phase possesses sufficient ductility. Combining Eq 3 and Eq 4, the following expression for q is obtained.

$$q = \frac{\sigma_\beta - \sigma_\alpha}{\varepsilon_a - \varepsilon_\beta} \qquad \text{(Eq 6)}$$

In principle, it is possible to obtain numerical values of q, from Eq 6, using properties of individual phases. However, two important questions arise.

- As the extent of the deformation is increased, the flow strength of the matrix increases. This may cause a continuous change in the q value during deformation.
- Equation 6 may be used for estimating the q value, provided it is previously known that the ductilities of the phases are exhausted at the ultimate tensile strength (UTS) of the composite.

As a first approximation, it may be assumed that q remains virtually unaltered during the deformation range of uniaxial tensile test. Assuming that the ductilities of both phases are exhausted just before the UTS of a composite is reached, the q

value may be estimated. This can be, however, the starting value only, and a more realistic value is achieved only by trial and error.

3. Three-Phase Steels

Among multiphase steels, three-phase steels containing ferrite, martensite, and retained austenite appear to be very promising, as already demonstrated.[5-7] In such materials, strain-induced transformation of retained austenite provides considerable leverage in bringing about an improvement in the strength-ductility combination.[5] Similar steel compositions, with or without small additions of alloying elements, may be processed to obtain varying amounts of austenite with differing stabilities to obtain different strength-ductility combinations. The choice of processing parameters will, therefore, depend upon end application.

The modeling of structure-property relations in these steels may be carried out in a manner similar to the preceding with certain modifications. These will be described in detail later in the text.

3.1 Experimental

In the present work, two steels with different chemistries (C-0.08, Mn-1.15, Si-1.3, and Cr-0.44 [Cr steel] and C-0.14, Mn-1.95, Si-0.6, and V-0.08 [V steel]) have been subjected to a two-stage thermal cycle so as to obtain varying amounts of ferritic, martensitic, and/or bainitic phases with retained austenite.[6,7] The amount and stability of retained austenite varies and depends upon the processing route.

Retained austenite was determined employing a Siemens D500 x-ray diffractometer, using a molybdenum target. Miller's equation,[8] given subsequently, was used to evaluate the amount of retained austenite.

$$\gamma_R = (1.4 I\gamma)/(I\alpha + 1.4 I\gamma)$$

where

$I\gamma$ is the average of the integrated intensities obtained from (220) and (311) planes of retained austenite, and

$I\alpha$ is the integrated intensity from the (211) plane of ferrite and martensite.

3.2 Mechanical Stability of Retained Austenite

Samples for Cr and V steels treated to retain a maximum of austenite were deformed under tension to estimate their mechanical stability. The magnitude of strain was varied from 0.05 to 0.2 in steps of 0.05, and the amount of nontransformed austenite, as before, was measured by x-ray diffractometry (Table 1).

The stability of retained austenite under differing conditions of deformation is compared in Fig. 1 on the normalized scale based on the data of Table 1 for the experimental steels. The difference in the rate of decrease in the retained austenite for the two steels can be explained by invoking a parameter, which can be defined in terms of austenite stability upon deformation.

3.3 Analysis

From Eq 3 and Eq 6, we obtain

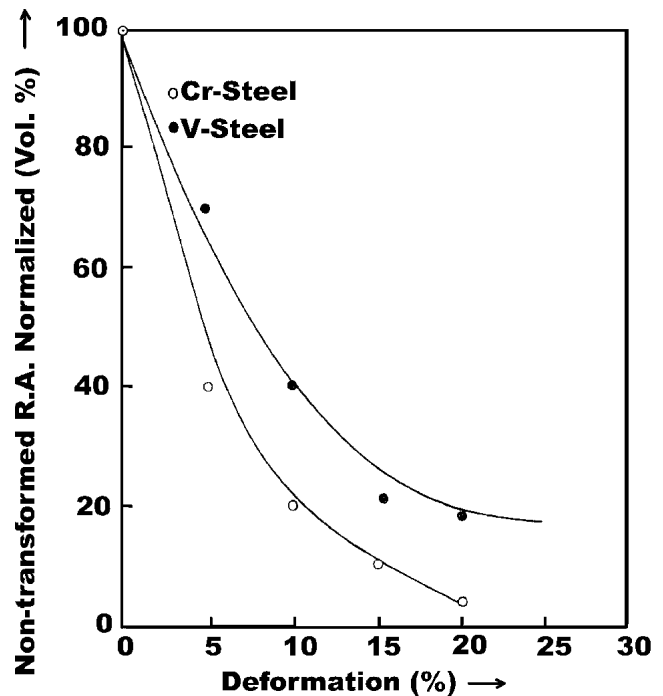$$\sigma_c = V_\alpha \sigma_\alpha + V_\beta \sigma_\beta \tag{Eq 7}$$

Fig. 1 Effect of tensile deformation on normalized volume percent of nontransformed retained austenite

Table 1 Change in Volume Fraction of Retained Austenite with Increasing Strain

	Volume Fraction of Retained Austenite	
Strain	Cr Steel	V Steel
0.00	15.1	10.2
0.05	6.3	7.2
0.10	3.1	4.1
0.15	1.5	2.0
0.20	0.5	1.8

Equations 5 through 7, although not amenable to analytical solution, can be solved numerically to obtain a stress-strain history of a two-phase material. This, however, necessitates prior knowledge of the stress-strain relationship of the component phases. Adopting the analysis as proposed by Kim,[3] the following relations have been used:

- Up to the strain, ε_0,

$$\sigma = A\varepsilon^2 + B\varepsilon + \sigma y \tag{Eq 8}$$

- Beyond the strain, ε_0,

$$\sigma = K\varepsilon^n \tag{Eq 9}$$

To assess the reasonableness of the model and assumptions made therein, a comparison was attempted between the computed and experimental stress-strain curves. Experimental curves were available from the two steels, that is, V and Cr steels (Fig. 2 and 3). These steels were processed under different conditions so as to obtain varying amounts of martensite and metastable-retained austenite, keeping the total of second-
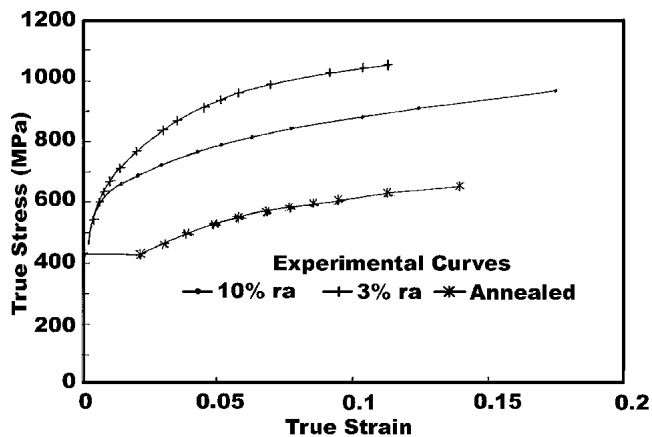
Fig. 2 Experimental stress-strain curves for V steel (ra, retained austenite)
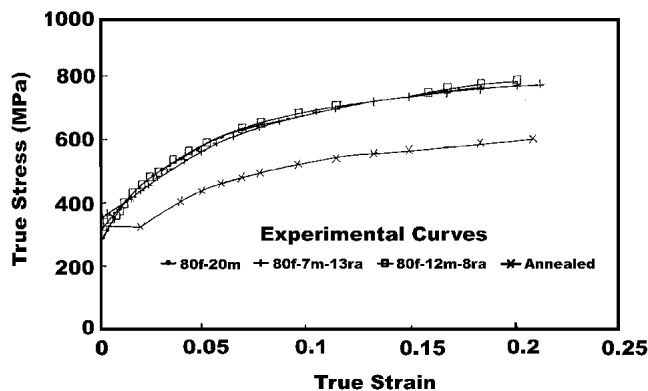

Fig. 3 Experimental stress-strain curves for Cr steel (f, ferrite; m, martensite; and ra, retained austenite)

ary phases at about 20% by volume fraction. It may be noted that the experimental curves that were actually determined for the matrix fit very well with the parabolic form of Eq 8. However, the two-stage approximation, outlined in Eq 7 through 9, became necessary to enable termination of numerical calculations at the end of the uniform elongation. This was achieved by using the classical condition that at the end of the uniform elongation, the work-hardening rate becomes equal to the stress at that point.

Parameters for soft-ferrite phase were estimated by conducting tensile tests on both Cr and V steels furnace cooled from austenitizing temperature. Figures 2 and 3 show the true stress-strain plots for V and Cr steels, respectively. The yield strength of the annealed Cr sample was found to be 320 MPa, whereas for V steel, it was 415 MPa. The microstructure of the samples consisted of ferrite and carbides. Although the presence of carbide precipitates may have somewhat altered the actual ferritic stress-strain curve, it is believed that this approximation is better than the use of literature data on other ferritic steels with different composition. The difficulty, however, arose in choosing the stress-strain relations for the martensitic component. The properties of martensite used in reference 9 yielded good results for Cr steel. For V steel, however, the use of these properties resulted in very low estimates, as compared to the

experimental results, for the yield strength and UTS of the composite. It was, therefore, decided to estimate the strength of martensite of V steel by considering the partitioning of carbon in martensite and use of carbon versus strength curves from reference 10.

The effect of retained austenite was included in the preceding calculations by using the following procedure.

1) Prior to strain-induced transformation, properties of austenite are presumed to be the same as that of the soft-ferritic matrix.
2) An exponential equation, $f_v = f_0 \exp(-p\varepsilon)$, has been used to describe the rate of strain-induced transformation, and the following expression has been used for the instability factor, p:

$$p = \left(\ln\frac{f_0}{f_v} \right) / \varepsilon \qquad \text{(Eq 10)}$$

where ε is the reference strain at which fraction, f_v, of the original austenite remains nontransformed.

The values of f_v and ε were estimated from Fig. 1, which depicts the change in the amount of retained austenite with deformation for both steel chemistries. The p values have been computed from Fig. 1 and redrawn on semilogorithimic plots for the two steels. They work out as 15 and 8 for Cr and V steels, respectively. The difference in the instability parameters of the two steels has its genesis in the shape and size distribution of retained-austenite particles.

Equation 10 implies a continuous transformation of retained austenite with increasing strain. In practice, however, this may not be the case. Because of heterogeneous chemistry and size distribution, individual particles of retained austenite will have different stabilities and, therefore, undergo martensitic transformation over a range of strains. The sequence of transformation thus may be discrete rather than continuous. These considerations could be incorporated into the calculations by choosing appropriate incremental-strain intervals at which calculations are made.

On the basis of the preceding model, numerical solutions have been obtained using the actual volume fractions of phases present in various samples, and typical solutions are shown in Fig. 2 through 5. Note that good agreement between the theoretical predictions and the experimental data, as shown in Fig. 4 and 5, was obtained by optimizing the value of the load-transfer factor, q, and by using an additional term, such as the one indicated in Eq 7. These two factors are interesting features of the present work and need some elaboration.

The choice of an appropriate q value for predicting the flow behavior of dual-phase steels is very important. Fischmeister and Karlsson[11] have determined the q value in an interesting manner from the ratio of the strains in the hard and soft phases, which were estimated from the microhardness data together with the flow curves of the two phases. The q value has also been determined by Tomota et al.[12] using field emission microscopy (FEM) analysis. They have associated the physical meaning of q with plastic relaxation because the process of plastic relaxation relieves the internal stresses by decreasing the effective plastic-strain differences between the hard and soft phases. In the present work, the initial q values were es-
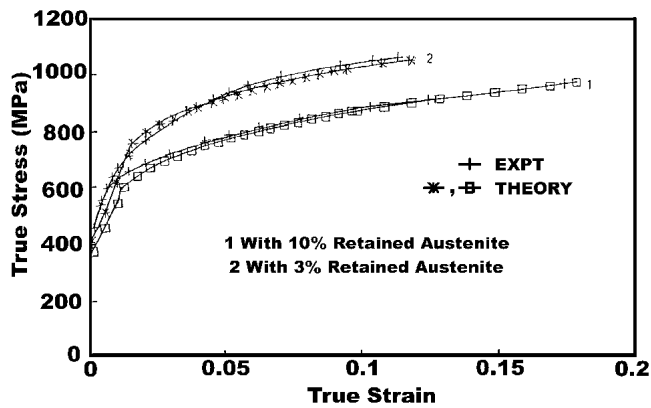
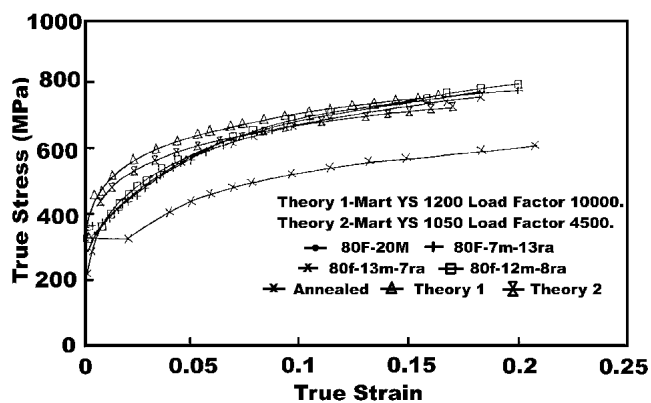**Fig. 4** Comparison between theoretically predicted and experimentally determined behavior (V steel)

Fig. 5 Comparison between theoretically predicted and experimentally determined behavior for Cr steel (f, ferrite; m, martensite; and ra, retained austenite)

timated from stress-strain curves of individual phases, but did not yield a reasonable agreement with the experimental data. The actual values used in the calculations were, therefore, obtained by trial and error and were found to be 10^4 for the Cr steel and 10^5 for the V steel. These values represent different levels of back stresses that are generated in the two steels investigated, and suggest that the range of 3,000 to 6,000 indicated by Goel et al.[1] is too narrow to apply to a variety of steels.

Despite the use of appropriate q values, the theoretical curves did not compare that well with the experimental curves, as shown in Fig. 5 for Cr steel. Although not much improvement was found possible in the case of Cr steel, the situation improved considerably by using an additional term for the V steel, as shown in Fig. 4. This additional term was obtained by increasing the effective volume of the hard phase by an amount equivalent to the square root of strain up to 10%. The physical significance of such an additional term is linked with the dislocation interface between the matrix and hard particle(s). As the deformation proceeds, dislocations continue to produce more dense networks at the interface until at a certain critical strain, say 10%; any additional dislocation escapes by a mechanism such as cross slip. Thus, the thickness of this interface increases continuously with strain, and it has a strength, which

is an average between that of the matrix and the hard particles. For the purpose of calculations, the actual volume of this interface is converted to an equivalent volume of a hypothetical interface with strength equivalent to that of the hard particle itself. The addition of this new term may also be justified by the principles of nonhomogeneous deformation, as discussed by Ashby,[13] who has shown that compatible deformation of a soft material containing a hard, less deformable second phase requires the generation of plastic-strain gradients (i.e., the gradients of geometrically necessary dislocations) within the deformable matrix. The results of Araki et al.[14] have also shown that the harder martensite would allow development of greater nonhomogeneous internal stress fields, which could more effectively restrict dislocation motion. The term considered to account for the "geometrically" required dislocations around the hard particles is proportional to $\sqrt{V_\beta}\varepsilon$, and inclusion of the term in the present modeling proved helpful in achieving results similar to that discussed earlier. Because the existence of small hard-phase particles has been observed in the V steel,[8] such an interpretation cannot be ruled out. Further investigations are required to determine the nature of this additional term and its physical significance.

4. Conclusions

A theoretical model has been attempted for ferrite-martensite dual-phase and ferrite-martensite-austenite multi-phase steels. The model matched well with the experimental results when the appropriate value of the load-transfer factor, q, was taken, depending on the stability of the retained austenite. The strength of the particle-particle interface appears to play a key role in these models, and additional terms linked with the dislocation interface between the matrix and the hard particles should be incorporated to achieve better approximation in the model.

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