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# Effect of pre-IC annealing treatments on the final microstructure and work hardening behavior of a dual-phase steel

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ABSTRACT

This paper investigates the relationship between the microstructure and the work hardening behavior of a dual-phase (DP) steel. Various DP microstructures were systematically produced by application of different pre-IC annealing heat treatments as well as changing the IC annealing temperature. It was found that various austenite nucleation sites such as grain boundaries, prior pearlite colonies, martensite particles and cementite particles, have different nucleation and growth effectiveness which significantly influences the microstructure after IC annealing. Following a quantitative analysis of all microstructures, the effect of microstructural parameters including martensite particle size, volume fraction as well as their spatial distribution and morphology, on the mechanical behavior of DP steels is examined by considering true work hardening rate, instantaneous work hardening rate and the dislocation annihilation factor from the Kocks–Mecking analysis. These analyses reveal that at constant ratios of volume fraction to size of martensite particles, there are significant differences in all the three work hardening parameters. It is proposed that these observations are due to the effects of morphology and spatial distribution of martensite particles. Furthermore, it was shown that the contribution of martensite particles to work hardening behavior, via geometrically necessary dislocations, is only significant at the early stages of deformation.



## 1. Introduction

There is a continuing trend in the automotive industry to move towards lighter, more fuel-efficient vehicles. To ensure competitiveness of ferrous alloys, new grades of Advanced High Strength Steels (AHSS) are being developed with a superior combination of strength and formability. The present research investigates dual-phase (DP) steels which contain non-ferritic phase (NFP) particles distributed in a ferrite matrix. In traditional DP steels, martensite is the only NFP. However, for more complicated processing routes, other NFPs such as bainite are also present. The standard processing of DP steels involves inter-critical (IC) annealing of cold-rolled ferrite-pearlite steel in the ferrite-austenite phase field. During this treatment, austenite nucleates at the interfaces between ferrite and cementite particles, either as individual particles or within pearlite colonies [1–3]. The carbides provide the elevated carbon content required for the growth of austenite. However, their effectiveness as austenite nucleation sites varies with their location within the microstructure. Earlier studies [4,5] have shown that pearlite is the first phase that is dissolved and replaced by austenite. This process is very rapid and occurs within a few seconds of heating to the IC temperature [5]. The effect of individual carbide particles on austenite nucleation, however, is more complex. Following pearlite colonies, grain boundary carbides are primary austenite nucleation sites, whereas grain interior carbides are less effective for austenite nucleation [5–7]. This difference can be attributed to the extra surface energy associated with the grain boundaries. It is therefore expected that the starting microstructure, i.e. prior to IC annealing, directly influences the final DP microstructure, a correlation that has been demonstrated in several studies [1,3,8–13].

DP steels exhibit continuous yielding, low yield-to-tensile strength ratios, large uniform elongations and very high initial work hardening rates. The continuous yielding of DP steels at low stresses is associated with the presence of mobile dislocations at ferrite/martensite interfaces as well as residual stresses produced by quenching from the IC annealing temperature [11,14–18]. On the other hand, the high initial work hardening rate is a direct consequence of the strain incompatibility between ferrite and martensite which results in load transfer between soft, deformable ferrite and hard, non-deformable martensite [10,11,18–20]. Furthermore, this strain incompatibility results in the introduction of additional hardening mechanisms in the ferrite matrix due to the development of back stresses and the production of geometrically necessary

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dislocations (GNDs) near particle interfaces [11,17,18,20–22]. However, as total plastic strain is increased, the effectiveness of these additional hardening mechanisms is reduced and subsequently the work hardening behavior of DP steels is determined by the balance between the accumulation of statistically stored dislocations [20,21,23–25] and dislocation annihilation due to dynamic recovery [26,27].

Microstructural parameters such as ferrite grain size as well as the volume fraction, size, morphology and spatial distribution of martensite particles are known to have a significant effect on the tensile behavior of DP steels [10,11,18,24,28–36]. The volume fraction of martensite, $f$, is found to directly affect the amount of load transfer between ferrite and martensite [10], the yield strength [11,12,21,22,29,32,33] as well as the work hardening rate due to back stresses [20,24,30] and GND hardening [9,11,20,31]. The mean size of martensite particles, $d$, is known to inversely affect the tensile properties, in particular work hardening due to GNDs [11,31]. Consequently, on the basis of Ashby's work hardening model for the deformation of plastically inhomogeneous alloys [37,38], work hardening models for DP steels typically consider the parameter $\sqrt{f/d}$ as the main microstructural parameter that controls the uniaxial tensile behavior [11,24,31]. In addition, the spatial distribution and morphology of martensite have also been shown to influence the yield strength [1,36], work hardening rate [8,10,23,24,35] and load transfer between ferrite and martensite [10]. The effects of DP steel microstructural parameters on dislocation annihilation effects by dynamic recovery have not been studied extensively. There are only a limited number of studies [26,27] that model dislocation annihilation effect using a Kocks-Mecking approach for DP steels [39–41] but these models are very general and do not incorporate the effects of microstructural parameters.

The majority of reported studies on the work hardening behavior of DP steels do not distinctly separate the effects of martensite morphology and spatial distribution from that of martensite size and volume fraction. Therefore, the objectives of the present study are to utilize pre-IC annealing heat treatments to produce distinct variations in the morphology and spatial distribution of martensite particles at a constant $\sqrt{f/d}$ ratio, and to investigate the effects of these microstructural parameters on the work hardening behavior.

## 2. Experimental procedure

### 2.1. Materials

The material used for this study was provided by US Steel Canada as cold-rolled, commercial DP780 grade, 0.95 mm thick sheet. The composition is given in Table 1 (DP780-CR). Blanks of size 100 mm by 20 mm were cut from the CR sheet with the long axis either along the rolling direction (RD) or the transverse (TD) direction. Only a limited number of blanks were prepared along TD. Therefore, all specimens are assumed to be RD, unless specified otherwise. In addition to the DP780-CR material, an interstitial free (IF) steel sheet was examined for comparison purposes. Its composition is also listed in Table 1.

### 2.2. Heat treatments

Two different "pre-treatments" were applied to the CR material in order to produce additional microstructural variants prior to the IC annealing process:

(a) *Austemper (AT)*: The aim of this treatment was to produce a bainitic structure. The processing involved full austenitization followed by an isothermal hold in the bainitic transformation region. Austenitization treatments were performed at 920 °C for 30 min in a Lindberg 54232 tube furnace under continuous argon flow to avoid oxidization of the specimen's surface. The end-to-end variation in the temperature along the length of the heat treating blanks was ±10 °C. The bainite hold step was carried out at 500 °C for 20 min in a salt bath of potassium nitrate and sodium nitrate mixture. The temperature gradient over the sample dimensions in the salt bath was determined to be ±2 °C. The cooling rate during the transfer of specimens from the austenitization furnace to the bainite hold salt bath was ~13 K/s. The bainite hold temperature was selected to be midway between the bainite start (575 °C) and martensite start (430 °C) temperatures calculated using Steven and Haynes [42] and Andrews [43] formulae, respectively. All specimens were water quenched to room temperature after the bainite hold treatment. The resulting microstructures will be referred to as AT.

(b) *Quench and Temper (QT)*: The aim of this treatment was to produce a tempered martensite structure which contains carbide particles with morphologies and distributions that are distinctly different from the AT microstructure. This involved a full austenitization treatment, a subsequent water quench to complete the martensitic transformation and finally a temper at 600 °C for 1 h in order to produce a fully tempered martensite microstructure. Similar to the procedure for the austenitization of AT specimens, both austenitization and tempering treatments of the QT samples were performed in the Lindberg 54232 tube furnace for 30 min under continuous argon flow. Three different austenitization temperatures of 920, 970 and 1020 °C were used in order to change the prior-austenite grain size. The resulting microstructures will be referred to as QT.

The final processing stage involved IC annealing of pre-treated specimens (CR, AT or QT) such that the final microstructures will be referred to as CR+IC, QT+IC and AT+IC, respectively. The QT microstructures are further categorized into Q2T, Q7T and Q12T based on the austenitizing temperature, i.e. 920, 970 and 1020 °C, respectively. IC annealing was performed at different temperatures of 720, 725, 730 and 735 °C, in order to produce microstructures with various volume fractions of martensite. IC annealing treatments were done in a salt bath with an average heating rate of ~19 K/s. The end-to-end variation in the temperature along the length of the heat treating blanks was ±2 °C. The IC annealing time began 40 s after immersing the specimens in the salt bath. This timing was chosen based on the average time required for specimens to reach the $Ac_1$ temperature, of 700 °C, determined using the Andrews' formula [43]. The majority of specimens were

**Table 1**
Chemical compositions of DP780-CR and IF steel sheets (in wt%).

| Steel | C | Mn | P | S | Si | Cu | Ni | Cr | Mo | N | V | B | Ti | Nb |
|---|---|---|---|---|---|---|---|---|---|---|---|---|---|---|
| DP780-CR | 0.09 | 2.1 | 0.012 | 0.006 | 0.02 | 0.03 | 0.01 | 0.26 | 0.29 | 0.004 | 0.001 | 0 | 0.001 | 0.002 |
| IF | 0.004 | 0.12 | 0.003 | 0.008 | 0.008 | 0.019 | 0.014 | 0.011 | 0.005 | 0.004 | 0.02 | 0.003 | 0.063 | 0.005 |