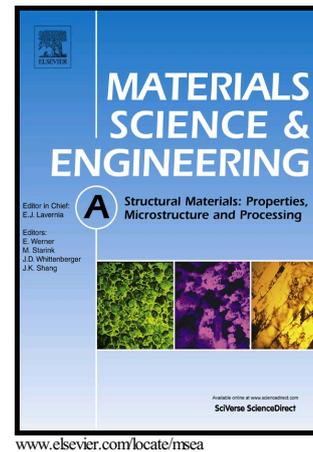


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# Extraordinary strength and ductility obtained in transformation-induced plasticity steel by slightly modifying its chemical composition

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## Abstract

Extraordinary mechanical properties were obtained in a low-carbon transformation-induced plasticity (TRIP) steel by modifying its chemical composition and employing thermo-mechanical processing. It was found that Cu addition to the conventional CMnSiAl TRIP steel could improve the thermal stability of high temperature austenite by decreasing the  $A_{c3}$  temperature and retarding the bainitic transformation, which led to the formation of retained austenite with higher volume fraction but lower carbon content. The combined effect of high strain hardening due to the TRIP effect, uniform strain distribution, higher volume fraction of thermal martensite, and nano-scale precipitation of  $\epsilon$ -Cu particles in the Cu-TRIP steel led to a high strength of 1 GPa and total elongation of approximately 50%. Moreover, a detailed scanning electron microscopy analysis showed that the copper had a noticeable effect on the morphology of the martensite phase created within the microstructure.

**Keyword:** TRIP Steel; DP Steel; Chemical Composition; Mechanical Properties.

## 1. Introduction

Zackay et al. introduced the transformation-induced plasticity (TRIP) phenomena in austenitic stainless steels for the first time [1]. In these grades of stainless steel, the austenite phase could be retained because of the presence of a high level of an expensive alloying elements such as Ni and Cr. Thereafter, numerous studies have been conducted to determine whether it is possible to design a new heat treatment cycle for low alloy steels that allows them to retain up to 20% austenite at room temperature with adequate stability. As a consequence of these studies, dual-phase (DP) steels were developed in the mid-70s. It was shown that a microstructure consisting of martensite (M) grains in a ferrite (F) matrix could lead to improved mechanical properties. This dual microstructure is generated by intercritical annealing (IA), i.e. heating between  $A_1$  and  $A_3$ , followed by water quenching. The austenite phase that is formed during the intercritical annealing is transformed into martensite by quenching [2]. Dual-phase steels mostly contain small amounts of retained austenite (RA) at room temperature. However, some studies have shown that this austenite did not improve the mechanical properties through the TRIP effect [3]. Although intercritical annealing and quenching (DP treatment) do not yield a large amount of retained austenite, these grades of steel still contain enough carbon, which is a very strong austenite stabilizer element, to induce adequate mechanical stability of the austenite at room temperature. Nonetheless, further carbon enrichment of the austenite is required in addition to the main

enrichment that occurs during the intercritical annealing. Such an extra enrichment can result from an inherent feature of the bainitic transformation [4-6]. This idea led to the development of multiphase (MP) CMnSi TRIP steels. The microstructure of TRIP steels is conventionally comprised of polygonal ferrite (PF), bainite, retained austenite (10–20%), and martensite [7, 8]. This special microstructure is obtained by performing a two-stage heat treatment. Similar to dual phase steels, the first stage consists of intercritical annealing, during which a part of the initial microstructure is transformed into austenite. In contrast to dual phase steels, which are directly quenched to room temperature, TRIP steels are rapidly cooled to an isothermal bainitic transformation (IBT) temperature (typically 350–450 °C). This particular heat treatment cycle is designed for the carbon enrichment of austenite, which stabilizes the retained austenite even at room temperature [5, 7, 8]. During the second stage, a portion of the intercritical austenite is transformed into bainite, whereas the remaining austenite is stabilized in such a way that it does not transform into martensite during quenching to room temperature. The stabilization of the austenite during the bainitic transformation is due to carbon rejection from bainitic ferrite into the residual austenite and to the inhibition of cementite precipitation from austenite when the steel contains enough silicon. It is well known that the displacive growth of bainitic ferrite is followed by the rejection of excess carbon from the bainitic ferrite into the surrounding residual austenite, where cementite precipitation occurs [9]. This cementite precipitation can be absolutely inhibited in the presence of silicon, because silicon has a very low solubility in the cementite phase [6, 7, 10]. Therefore, the bainite transformation leads to the formation of a mixture of carbide-free bainitic ferrite and carbon-rich residual austenite. The high carbon content in this residual austenite brings the  $M_s$  temperature below room temperature. Therefore, conventional TRIP steels always contain 1–2% silicon to prevent cementite precipitation. In addition, Si contributes to the strengthening of ferrite by solid solution hardening. Unfortunately, a high Si content in TRIP steels can result in the formation of a very strong oxide layer, which can prevent the formation of the inhibition layer during the hot-dip galvanizing process [11]. The partial or complete replacement of the Si with Al [12, 13] or P [14, 15] has been considered to address this issue. However, because of the more potent solid solution strengthening effect of Si and its greater ability to inhibit carbide precipitation compared with Al and P, the complete replacement of Si with Al or P has not been recommended. Thus, new CMnSiAlP TRIP steels have been developed in recent years [15].

Reviewing the literature shows that, in recent years, great emphasis has been placed on improving the mechanical properties of TRIP steels through modifying the heat treatment cycle and adding the alloying elements to conventional CMnSiAl steels. Ding et al. proposed a modified heat treatment cycle to increase the volume fraction of retained austenite in low-alloyed steels [16]. By intentionally retaining a small amount of interlath-retained austenite before annealing at a relatively high temperature, new intercritical austenite was encouraged to pursue an acicular morphology between laths. Therefore, the formation of blocky austenite, which usually exists during the intercritical annealing of the martensitic microstructure, was suppressed, leading to an increase in the RA content and improved mechanical properties [16].

Replacing the soft ferrite matrix by either hard martensite or carbide-free bainite through modifying the conventional two-step heat treatment of TRIP steels is considered to be another method to enhance the

mechanical properties of TRIP steels. For example, Speer et al. [17] proposed a new quenching and partitioning (Q&P) process to obtain a hard martensitic matrix with a controlled amount of retained austenite. The Q&P process consists of full austenization, followed by rapid cooling to an intercritical region between the martensite start ( $M_s$ ) and finish ( $M_f$ ) temperatures, and subsequent annealing at the quenching temperature or above to facilitate the partitioning of carbon from martensite to RA [17].

Zhou et al. developed a low-carbon Cu-bearing Nb-microalloyed TRIP steel [18]. By combining the TRIP effect and precipitation strengthening, a TRIP steel with a 29% volume fraction of retained austenite, yield strength greater than 700 MPa, total elongation of 35%, and uniform elongation of 22.8% was produced. They argued that the precipitation of nano-sized Nb precipitates inhibited the dislocation recovery by the pinning effect and the smaller Nb(C,N) with a size smaller than 10 nm combined with the 20 nm  $\epsilon$ -Cu precipitate formed by tempering significantly strengthened the matrix. Moreover, the use of copper as an austenite-stabilizing element is expected to stabilize austenite. However, few studies have considered the stabilization of austenite via the addition of copper.

Historically, copper has often been added to steel in small amounts to increase its corrosion resistance. At a level of ~1%, copper is the most effective element at inhibiting corrosion in steels [19, 20]. In addition, copper can decrease the transition temperatures ( $A_3$ ,  $A_1$ ,  $B_s$ ,  $M_s$ ) of steels [20]. Because copper has limited solubility in ferrite, it could cause a precipitation strengthening effect by forming FCC  $\epsilon$ -Cu precipitates through aging at temperatures of 500–700 °C. A copper addition of 1.5% has been found to slow the decomposition of austenite to ferrite by reducing the rates of ferrite nucleation and growth [21]. This implies the possibility of copper having a profound influence on TRIP steels. Slowing the ferrite transformation might allow easier production of austenite-enriching bainite, aiding the TRIP effect. On the other hand, it could potentially suppress all of the diffusional processes, poisoning the TRIP steel, and rendering it a dual phase ferrite-martensite steel by preventing the enrichment of the austenite during the isothermal bainite transformation [21]. The purpose of this paper is to examine the effects of copper addition on microstructure and mechanical properties of the conventional CMnSiAl TRIP and dual phase steels. The result of the present study will show that the copper can make a vital contribution to a future TRIP alloying scheme.

## 2. Materials and methods

The chemical compositions of the steels used in this study are listed in Table 1.

Table 1. Chemical compositions of produced steels.

Steel Code	%C	%Mn	%Al	%Si	%Cu	%Cr	%S	%P
A	0.23	1.65	1.5	0.52	0.02	0.01	0.006	0.018
B	0.21	1.7	1.6	0.52	2.43	0.01	0.06	0.017

The composition of steel "A" is typical of a conventional CMnSiAl TRIP multiphase steel with a copper content of 0.02 wt%. Steel "B" is a novel TRIP steel designed in this work with a copper content of 2.43 wt%. The steels were prepared by vacuum arc remelting (VAR) and cast into ingots with a thickness of ~7 mm.

The samples were then homogenized at 900 °C for 2 h and hot rolled in multiple passes to produce 2 mm strips. The strips were finally cold rolled to a thickness of 0.9 mm. Two heat treatment schedules were employed for the steels to produce dual-phase (DP) and multi-phase (MP) TRIP steels. These heat treatment conditions are represented schematically in Figure 1.

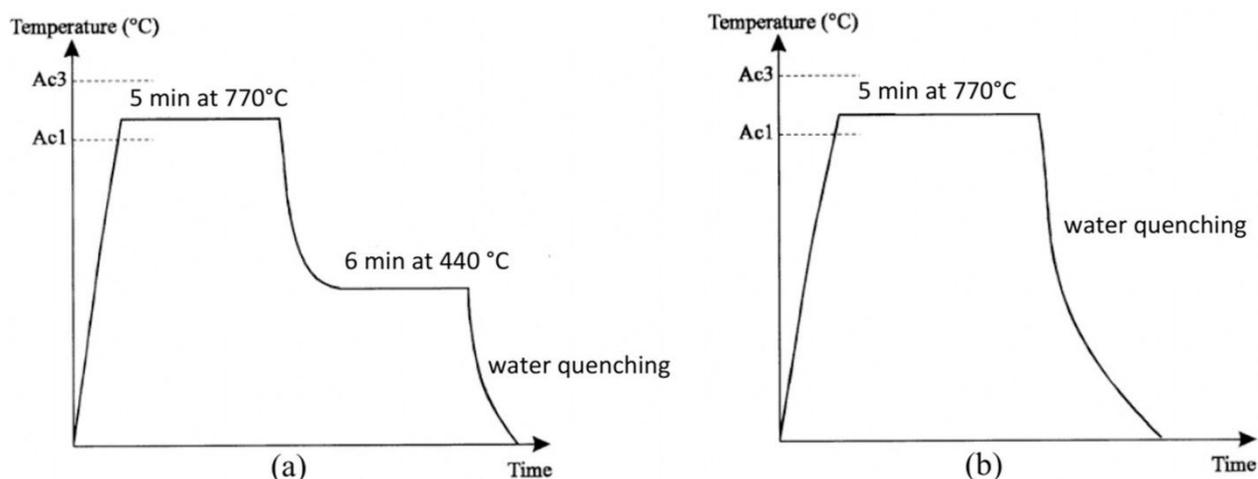


Figure 1. Schematic of heat treatment cycles used to obtain (a) multiphase (MP) TRIP and (b) dual phase (DP) steels.

The samples were first intercritically annealed at 770 °C for 5 min. The samples used to obtain the TRIP steel were then transferred to a salt-bath furnace for the bainitic transformation. Isothermal bainitic transformation (IBT) was conducted at 440 °C for 6 min. Moreover, to produce DP steels, the specimens were immediately water quenched after annealing. All of the applied treatments are summarized in Table 2.

Table 2. Treatments applied for production of DP and TRIP steels.

Specimen	Abbreviation	Applied Treatment
Low Cu DP steel	O-DP	5 min at 770 °C + water quenching
High Cu DP steel	Cu-DP	5 min at 770 °C + water quenching
Low Cu TRIP steel	O-TRIP	5 min at 770 °C + 6 min at 440 °C + water quenching
High Cu TRIP steel	Cu-TRIP	5 min at 770 °C + 6 min at 440 °C + water quenching

The samples for microstructural analyses were prepared using a standard metallographic method and etched with 2% Nital. The specimens were then examined using scanning electron microscope (SEM).

The volume fraction of retained austenite was measured by X-ray diffractometry (XRD) from the diffraction intensity ratio of (2 0 0), (2 2 0), and (3 1 1) in the FCC structure and that of (2 0 0), (2 1 1), and (2 2 0) in the BCC structure [22].

In addition, the carbon content of RA was determined from the austenite lattice parameter ( $a_{\gamma}$ ) according to the austenite peak positions in the XRD pattern. A combination of two approaches (Dyson and Holmes [23] and Ruhl and Cohen [24]) was used to compensate for the effect of manganese and aluminum additions on the

lattice parameter. Taking this into account, the retained austenite lattice parameter could be calculated from Equation (1) [25]:

$$a_{\gamma} = 3.572 + 0.0012 (\text{wt\% Mn}) - 0.00157 (\text{wt\% Si}) + 0.0056 (\text{wt\% Al}) + 0.033 (\text{wt\% C}) \quad (1)$$

where the values for manganese, silicon, and aluminum are obtained from the chemical composition of the steel (Table 1).  $a_{\gamma}$  is also determined from the position of the maximum of the three austenite reflections using Cohen's method [24].

Tensile test specimens were prepared as presented in Figure 2. Tensile tests were carried out at room temperature using a tensile test machine with a crosshead speed of 1 mm/min. Different tensile parameters were calculated, including the yield strength (YS), ultimate tensile strength (UTS), total elongation (TEL), uniform elongation (UEL), and product of the total elongation and tensile strength (TEL×UTS), as an index of the ability of steel to absorb energy.

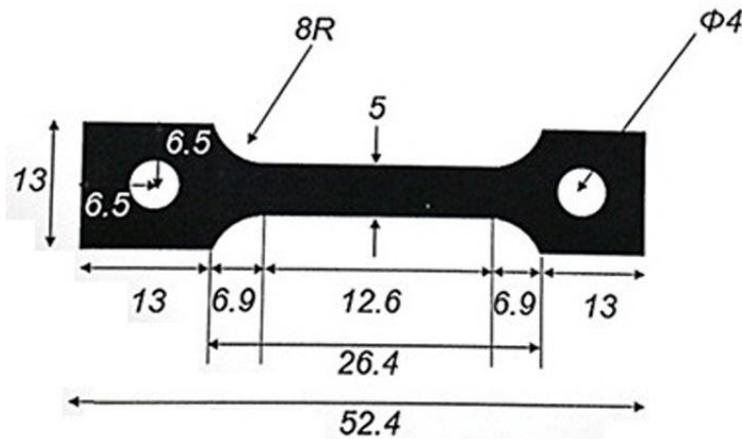


Figure 2. Engineering drawing of sub-sized uniaxial tensile specimen (all dimensions are in mm).

### 3. Results and discussion

Figure 3 shows the SEM images of the four heat-treated steels, previously described in Table 2. In the SEM micrograph of DP steels, Dark phase is ferrite and bright regions are martensite (Figures 3a–b). Moreover, austenite and bainite grains appear bright like martensite in the microstructure of the TRIP steel (Figures. 3c–d) [26]. Comparing Figures 3a and 3b clearly shows that, the DP steel that included 2.43% Cu (Cu-DP steel) incorporated a higher volume fraction of martensite than the specimen without Cu in its composition (O-DP steel). The martensite volume fractions of the studied DP steels were measured using image processing and found to be 66% and 41% for the Cu-DP and O-DP specimens, respectively. Supposing that all the austenite transforms to martensite in the DP steels, it is seen that on increasing the copper content up to 2.4 wt%, the kinetics of the austenite formation and growth increased significantly, up to approximately 60% in Cu-DP compared to that in the O-DP steel. In other words, a 2.4 wt% increase in the copper content increased the austenite stability, in the intercritical region, up to 60%.

The same behavior could be observed in the TRIP steels. The volume fraction of the bright regions in Figures 3c–d were calculated as 60% and 38% for the Cu-TRIP and O-TRIP steels, respectively. These values are

mainly the volume fraction of the austenite phase at the end of the IA stage, part of which subsequently transformed into bainite or martensite during the IBT stage and final water quenching. Again, Cu addition significantly enhanced the austenite stability in the intercritical region by increasing the volume fraction of austenite by about 60% at the end of the IA stage.

Using the Park formula that was specially developed for estimating the  $A_{c3}$  temperature in TRIP steels [27], the  $A_{c3}$  temperature in Cu-TRIP is found to be approximately 40 °C lower than that of O-TRIP. Thus, it is expected that, for the same intercritical annealing temperature for the two steels, the austenite fraction at the end of the IA stage would be greater for the Cu-TRIP steel, which is in good agreement with the present measurements.

Furthermore, considering the carbon content of ferrite as 0.02% and the volume fraction of austenite in the Cu-TRIP and O-TRIP steels as 60% and 38%, respectively, it can be shown that the carbon content of austenite at the end of the IA stage would be 0.33% and 0.57% for Cu-TRIP and O-TRIP steels, respectively. Thus, at the end of the IA stage, the Cu-TRIP steel would have a higher volume fraction of austenite with lower carbon content as compared with the O-TRIP steel.

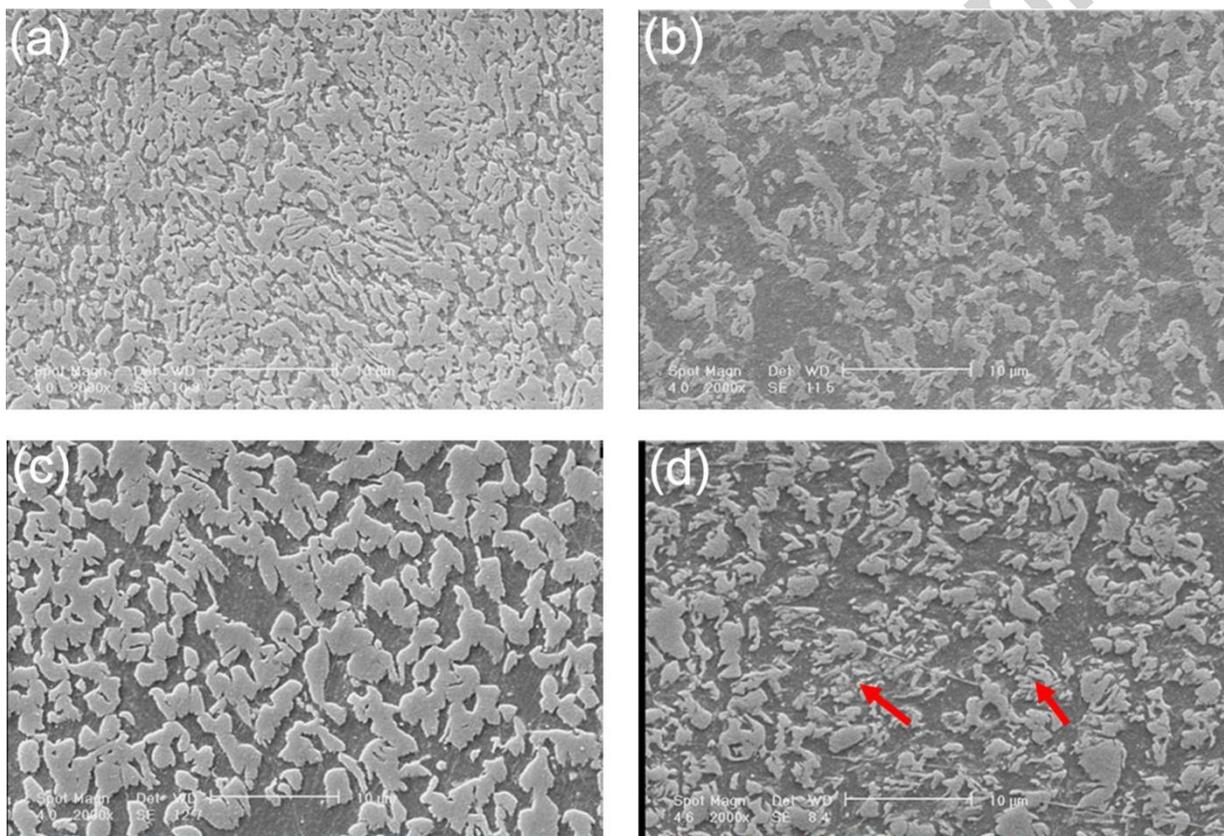


Figure 3. SEM micrographs of steels studied: (a) Cu-DP, (b) O-DP, (c) Cu-TRIP, and (d) O-TRIP. The red arrows indicate bainitic regions.

The advantages of increasing the austenite fraction with low carbon content in the intercritical region can be listed as follows:

- Increasing the final martensite volume fraction, after quenching in water, produces a higher strength

- Increasing the austenite volume fraction leads to a lower average carbon content for the resulting martensite [28], and thus a more ductile martensite is obtained.
- Decreasing the ferrite grain size makes it possible to obtain a more uniform strain distribution during deformation [29, 30].
- A greater stability for the austenite would result in an increase in the probability of austenite remaining at room temperature, and thus a greater TRIP effect would be expected.

Another observation seen in the figures is that the martensite grains in the steels that incorporated Cu have more rounded edges and their morphologies include more convex outer perimeters. In contrast, the martensite grain morphologies of the O-TRIP and O-DP steels are more fibrous.

As mentioned earlier, the non-ferritic phases, namely bainite, retained austenite, and martensite, appear as light contrast in the SEM micrographs of the TRIP steels [26], because of which it is impossible to distinguish these phases from each other only by observing the metallographic microstructures such as those presented in Figures 3(c) and (d). Thus, an XRD analysis was employed to measure the retained austenite content. Figure 4 shows the typical XRD patterns of the investigated steels. It was found that the Cu-TRIP steel incorporated about 21% of the retained austenite, while the O-TRIP steel contained 17% of the retained austenite, i.e., approximately 20% increase in the retained austenite fraction. This demonstrates the significant effect of Cu addition on the improvement of austenite stability up to room temperature.

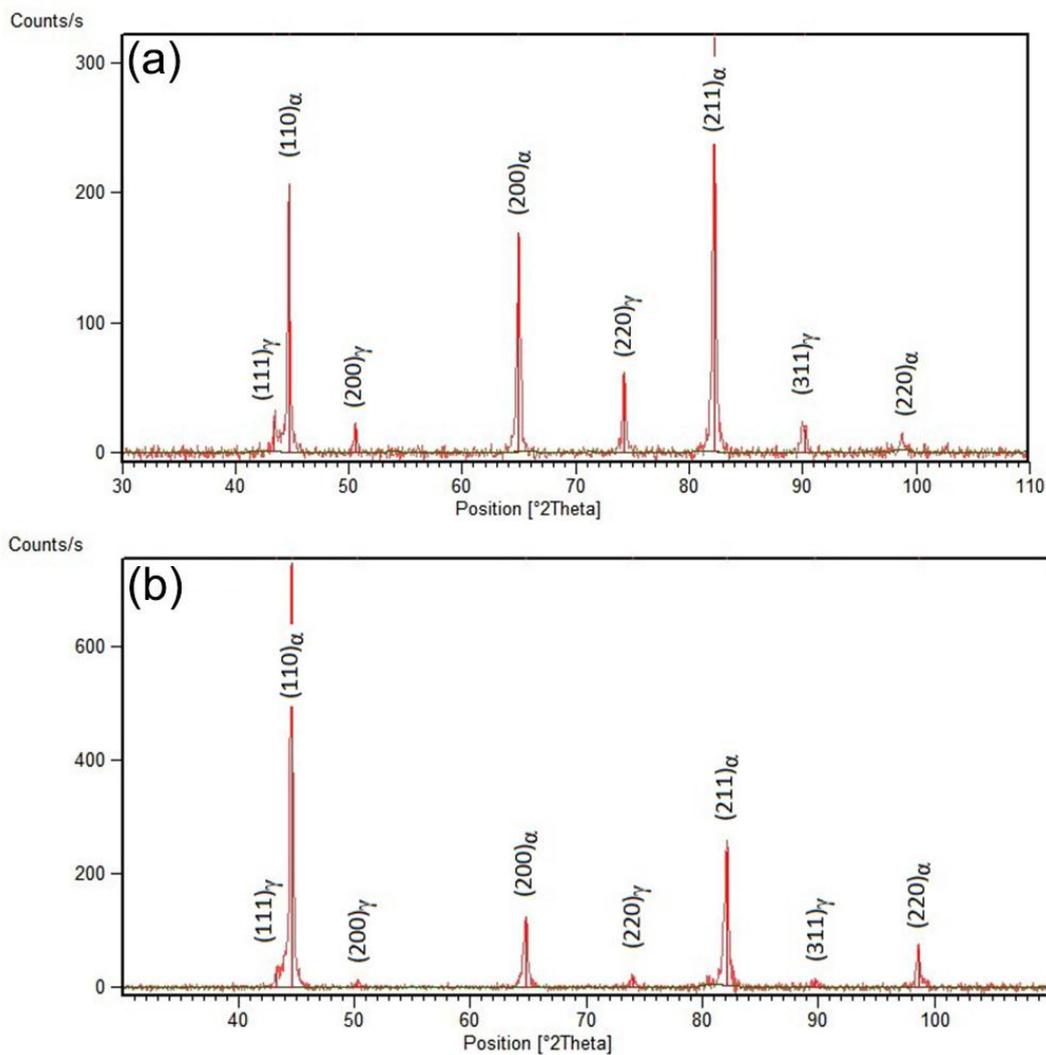


Figure 4. XRD maps of (a) Cu-TRIP and (b) O-TRIP steels.

Employing equation (1), the carbon content of the retained austenite for O-TRIP and Cu-TRIP steels were measured as 1.25 and 0.64%, respectively. The lower carbon content of the retained austenite in the Cu-TRIP steel can be attributed to the following reasons. Firstly, as mentioned earlier, the austenite phase in the Cu-TRIP steel has a lower carbon content as compared with that in the O-TRIP steel at the end of the IA stage (0.33% C for Cu-TRIP and 0.57% C for O-TRIP). Therefore, assuming that during the subsequent IBT stage carbon partitions in austenite to the same level for the two TRIP steels, the austenite in the Cu-TRIP steel is expected to have a lower carbon content as compared with that in the O-TRIP at the end of the IBT stage. However, Wang et al. [31] have recently shown that the main effect of copper in TRIP steels is to increase the stability of austenite phase by retarding the transformation of austenite to bainite during the IBT stage. This implies that the partitioning of carbon in austenite during the IBT stage is less effective in the presence of Cu, and therefore, the carbon enrichment of RA in Cu-TRIP is lower than that of RA in O-TRIP. In other words, the austenite phase in the Cu-TRIP steel not only possesses a lower carbon content at the end of the IA stage, but also during the subsequent IBT stage due to delayed bainitic transformation, lower amount of carbon partitioned into RA. Thus, the RA in the Cu-TRIP steel has a lower carbon content as compared with that in

As shown in Figure 3d, the microstructure of the O-TRIP steel contains some bainitic regions (indicated by red arrows), while those can hardly be distinguished in the Cu-TRIP steel. This phenomenon could be attributed to the effect of copper on retarding the austenite to bainite transformation.

Considering  $f_{RA}$  as the volume fraction of retained austenite and  $C_{RA}$  as its carbon content, the product of  $f_{RA} \times C_{RA}$  would be a measure of the amount of total carbon that is present in the RA in the final microstructure [32]. This value was calculated to be about 0.21% for the O-TRIP steel, which is very close to the average carbon content of steel (0.23%) (steel code A in Table 1). This indicates that almost all the carbon content of the O-TRIP steel are concentrated in the RA phase, and the volume fraction of other high-carbon phases such as martensite (if present) is very low. On the other hand, the product of  $f_{RA} \times C_{RA}$  for the Cu-TRIP steel was measured to be about 0.13%, which is far lower than the average carbon content of the steel, i.e., 0.21% (steel code B in Table 1). Therefore, in the Cu-TRIP steel, only about 65% of the total carbon was present in the RA and other phases with high carbon content such as martensite were also present in Cu-TRIP.

Using the empirical equation proposed by Mahieu for estimating the martensite start temperature of low-alloy TRIP steels [33], the  $M_s$  temperature of RA in Cu-TRIP and O-TRIP was calculated as 260 °C and 1 °C, respectively. To estimate the  $M_s$  temperature, it was assumed that only carbon completely partitioned between  $\alpha$  and  $\gamma$  during the process [34].

From these calculations, it can be concluded that, for the Cu-TRIP steel, a significant amount of austenite transformed into thermal martensite after the IBT stage. This conclusion is rational due to the low carbon content of RA in the Cu-TRIP steel (0.64% C), and thus higher  $M_s$  temperature, as compared with the O-TRIP steel (1.25% C). In contrast, no thermal martensite was formed in the O-TRIP steel after the IBT stage since the  $M_s$  temperature was far below the room temperature. Thus, almost all the carbon content of the steel remained in RA. This finding was previously predicted from the very low difference between the  $f_{RA} \times C_{RA}$  value and average carbon content of the steel.

Therefore, the final microstructure of Cu-TRIP steel (Figure 3c) comprises ferrite, RA, thermal martensite, and minor bainite, while that of the O-TRIP steel comprises bainitic phase with negligible thermal martensite (Figure 3d). Comparing the volume fraction of austenite at the end of the IA stage (38%) with that in the final microstructure (17%), it was seen that about 20% bainitic phase was present in the microstructure of the O-TRIP steel.

Uniaxial tensile testing was employed to examine the mechanical properties and deformation behavior. The engineering stress–strain diagrams of the steels studied are presented in Figure 5a. Moreover, the important mechanical properties are summarized in Table 3 and compared in Figures 5b–d. It is seen that the addition of approximately 2.4% Cu, or a 60% increase in the martensite volume fraction, for the Cu-DP steel leads to a 53% reduction in the TEL value, along with a 16% increase in the UTS level compared with the O-DP steel. On the other hand, in the TRIP steels, increasing the Cu alloying element up to approximately 2.4% led to an increase in the UTS level to 30%, while the reduction in the TEL value was not significant. This could be attributed to the its strain hardening behavior, which would be explained later, as well as the uniform strain

distribution obtained in the Cu-TRIP steel due to its higher retained austenite fraction, higher volume fraction of thermal martensite, and nano-scale precipitation of  $\epsilon$ -Cu particles. The combined effects of these factors led to an extraordinary strength of up to 1 GPa and total elongation of approximately 50% for the Cu-TRIP steel.

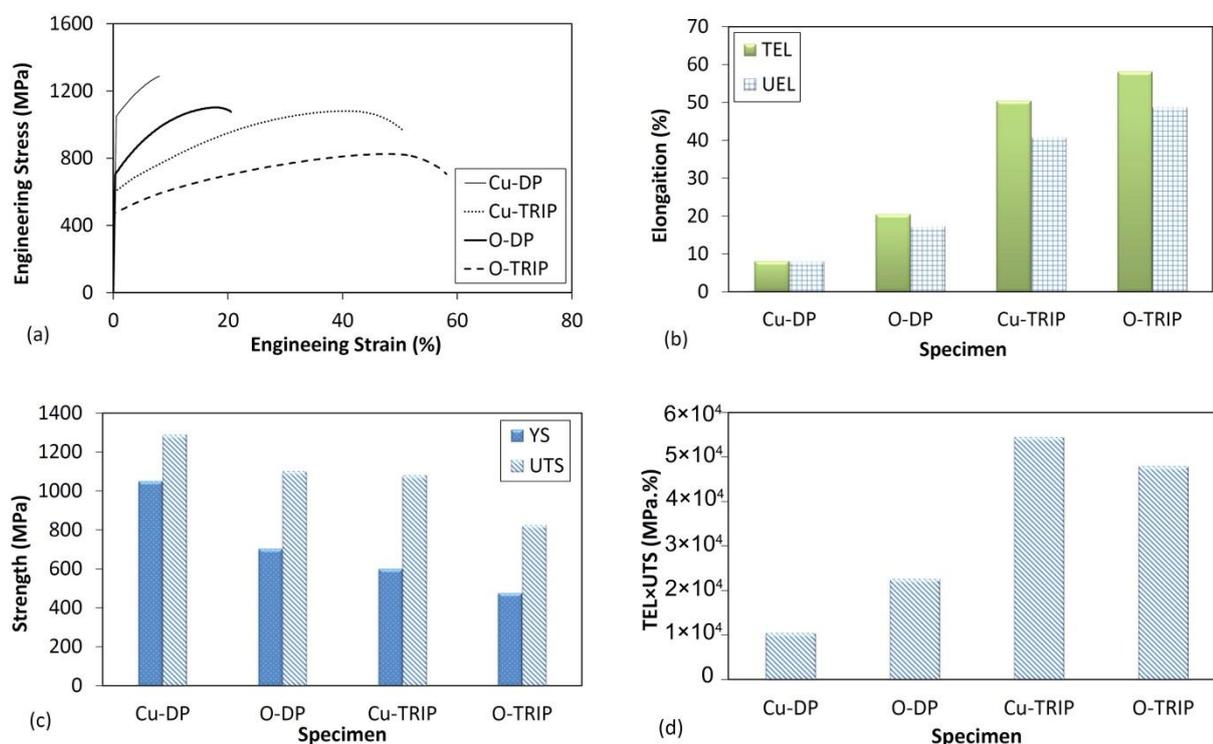


Figure 5. Comparison of tensile mechanical properties of steels studied: (a) engineering stress–strain curve, (b) elongation, (c) strength, and (d) fracture energies of DP and TRIP steels.

It is seen that the Cu-DP steel had a YS value that was approximately 33% higher than that of the O-DP steel. However, this increase for the TRIP steel was approximately 21%. This could be attributed to the fact that most of the dissolved Cu in the DP steel was concentrated in the martensite grains [20, 21]. Thus, in the Cu-DP steel, stronger martensite grains would lead to an increase in the YS level. However, in the Cu-TRIP steel, because of the higher amount of martensite plus retained austenite within the microstructure, the average dissolved Cu in the martensite grains would be lower than that expected in the Cu-DP steel. Thus, the lower strength martensite grains in the Cu-TRIP steel would be the reason for the smaller effect of the Cu addition on the YS value of the TRIP steels compared to the DP steels. The comparison between the yield strengths of the DP and TRIP steels indicated lower values for the latter (i.e., O-TRIP and Cu-TRIP steels). For example the yield strength of the Cu-DP sample was 1050 MPa, while for the Cu-TRIP steel which had the same composition but has an additional austempering stage, the yield strength decreased to 600 MPa (a reduction of ~43%).

Considering the value of  $UTS \times TEL$  as an indication of the required energy to materials fracture, it can be seen from Table 3 and Figure 5(d) that the Cu-TRIP steel involved with about 12% higher fracture energy than the O-TRIP steel. It is also observable that the Cu-TRIP steel involved with about 80% higher fracture energy than

Table 3. Important tensile properties of investigated steels.

Steel	YS (MPa)	UTS (MPa)	TEL (%)	UEL (%)	UTS×TEL (MPa.%)
Cu-DP	1050	1288	8.1	8.1	10433
O-DP	703	1102	20.5	17.4	22591
O-TRIP	475	825	58.1	48.7	47933
Cu-TRIP	600	1081	50.4	40.8	54482

Table 3 and Figure 5 also show that irrespective of the steel composition, the TRIP cycle definitely causes a larger TEL% in comparison with the DP cycle. This could be deduced from comparing the total elongation percentages of the O-DP/Cu-DP with O-TRIP/Cu-TRIP steels, which showed improvements of 180% and 520%, respectively. The exceptional TEL values of TRIP steels are directly related to the TRIP effect. The gradual transformation of austenite to martensite during the tension test caused a localized increase in the strain hardening coefficient that delayed the onset of necking. This led to the high elongation without compromising the strength level.

The variation of the strain hardening rate (SHR), defined as  $d\sigma/d\varepsilon$ , with the true strain for the four steels studied is presented in Figure 6a. It is obvious that the DP steels have a considerably higher SHR value compared to the TRIP steels. It is seen that the variation of SHRs are almost the same for both DP steels at a low strains. However, the decrease in slope for Cu-DP is sharper than that for the O-DP steel at high strain levels. Moreover, it is seen that the Cu-TRIP steel has an approximately 50% higher SHR values than the O-TRIP steel. As presented in the previous work of the authors [30], the decrease in the SHR value of a DP steel could be due to different phenomena, including 1) the grain rotation, 2) formation of substructure, 3) damage evolution, and 4) plastic deformation of the harder incorporated phase.

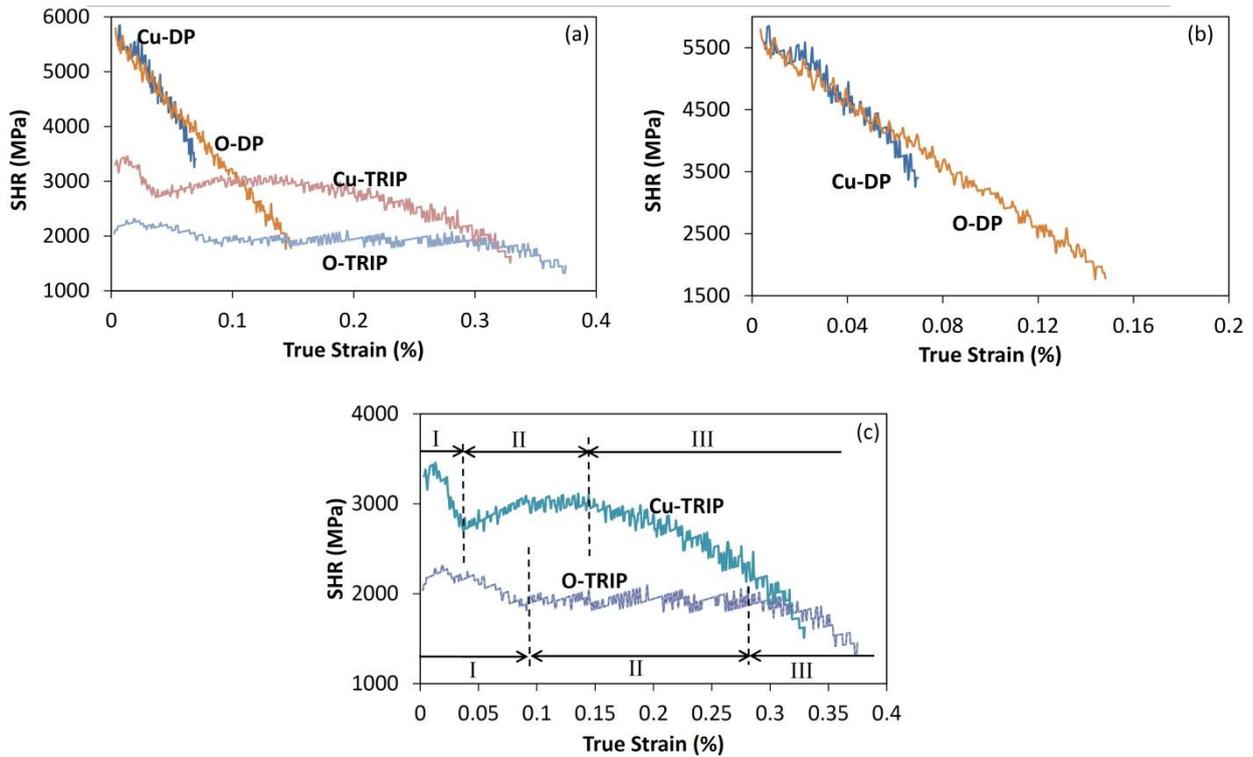


Figure 6. (a–c) Strain hardening rates of the steels studied.

To better distinguish the deformation behaviors of the DP and TRIP steels, the SHR diagrams of the steels studied were considered in more detail, as shown in Figures 6b–c. It is seen that in the O-DP steel, the SHR values decreased uniformly and with a constant slope. The SHR in the Cu-DP steel had almost the same behavior. However, in the TRIP steels, three strain hardening stages were observed:

**Stage I:** This stage is associated with the plastic deformation of ferrite, as well as the elastic-plastic deformation of martensite and TRIP transformation. In this stage, it was supposed that the dislocation pile-ups were annihilated by the initiation of the martensite plastic deformation. Moreover, the strain-induced transformation of austenite to martensite would accelerate the strain relaxation within the microstructure, leading to the subsequent decrease of the SHR.

**Stage II:** This stage is associated with the plastic deformation of ferrite and the elastic-plastic deformation of martensite. The interaction of different slip systems and creation of dislocation tangles, as well as dislocation annihilation by relaxation mechanisms [30], was supposed to result in a fairly constant change in SHR. Moreover, in TRIP steels, the strain-induced created martensite phase would act as a ban against the dislocation motion. This would cause an increase in the SHR in stage II, as seen in Figure 6c.

**Stage III:** This stage is associated with the plastic deformation of ferrite and martensite. The activation of different dislocation annihilation mechanisms such as the plastic deformation of martensite [35], substructure formation [30], void creation [36], and grain rotation [30] would accelerate the decrease of SHR in this stage.

Now, the question is why stage II for the Cu-TRIP steel is shorter and has a higher, increasing slope than that for the O-DP steel. In the microstructural examination of both TRIP steels in the previous section, it was

shown that the Cu-TRIP steel incorporated 19% more retained austenite. Moreover, the average martensite grain size was larger than that in the O-TRIP steel. The higher martensite volume fraction would act as a larger constraint against dislocation motion. Thus, a greater increase in SHR would be expected in the Cu-TRIP compared to the O-TRIP steel. Moreover, it could be said that in the microstructure incorporating the higher number of and larger martensite grains, the relaxation mechanisms initiating stage III would be activated at lower strains. Therefore, for the Cu-TRIP steel, the width of stage II is lower than that for the O-TRIP steel. This could be justified by the following mechanisms:

- A higher martensite volume fraction results in a lower average carbon content and thus a lower flow stress.
- A larger amount of martensite results in a larger number of void nucleation sites and higher probability of relaxation by damage creation.

#### 4. Conclusions

CMnSiAl and CMnCuSiAl steels were produced using vacuum melting. DP and TRIP steels were produced from these steels by employing homogenizing, intercritical annealing, austempering, and quenching treatments. Mechanical tensile testing and detailed microstructural examinations were used to study the effect of copper on the mechanical properties of the DP and TRIP steels. On the whole, the followings conclusions can be drawn:

- Increasing the copper content promotes the formation of a coarser and more rounded austenite morphology.
- The addition of 2.43 wt% Cu increased the austenite content up to 60% before the start of the IBT stage. In addition, the volume fraction of the retained austenite in the final microstructure increased by about 20%.
- The combined effect of copper in decreasing the  $A_{c3}$  temperature and retarding the bainite transformation during the IBT treatment resulted in a higher volume fraction of RA with lower carbon content.
- A significant elongation of approximately 50% and a tensile strength of 1 GPa were obtained when copper was added to the TRIP steel (Cu-TRIP).

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