EFFECT OF COPPER ON PROPERTIES OF FINE-GRAINED LOW-CARBON BORON STEEL

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Abstract

Low carbon steels are used in variety of structural states. One of the most used is fine grained dual or multi-phase structure with prevailing ferrite. Martensite/austenite islands forms rest of the structure. At least two of the phases - pearlite, bainite, martensite and rest austenite - occur in multi-phase steels. Main strengthening mechanisms are grain refinement and the hard phase content. The mostly ferritic soft matrix ensures sufficient plasticity. The constraining phenomena is cohesion of soft ferrite and the hard phase for the steel performance. Voids open upon deformation at the boundary between these constituents and lead to fracture. The difference in strength can be reduced by softening of the hard phase or by strengthening the ferrite. Copper alloying has possibility to do ferrite strengthening by precipitation. The experimental steel was subjected to controlled rolling to achieve dual- or multi-phase structure. Deformation in intercritical region resulted in fine ferritic matrix. Samples were subsequently quenched in water to transform remaining austenite into hard martensite or bainite. The delay between rolling and quenching gave opportunity for Cu to precipitate in freshly-formed ferrite and strengthen it. This approach leaves opportunity to gain precipitation strengthening in soft phase with no loss in hard phase strength. This is different from conventional quenching and tempering approach. The tempering acts also as precipitation hardening, but the tempering decreases strength of martensite in far higher rate than Cu precipitation can strengthen the ferritic matrix.

Keywords: Steel, Copper precipitation, Controlled rolling, Dual steel

1. INTRODUCTION

Copper is almost insoluble in ferrite at room temperature [1]. The solubility rises with increasing temperature and reaches maximum value 2 % at 850 °C (all element contents are given in mass %). The austenite has Cu solubility 2.5 % at that temperature. Therefore, there is a possibility for precipitation hardening - by Cu dissolution in ferrite at high temperature, rapid cooling into region of supersaturated Cu solution in ferrite and precipitation of Cu in form of nanometric particles. Structure refinement, which is essential for production of high-strength steels, is often ensured bz thermomechanical processing [2].

Controlled rolling is a production process which offers an opportunity to exploit copper precipitation phenomena without adding any processing stage to the steel sheet production. This process is already used to produce high strength steel sheets with ferritic-martensitic microstructure [3]. The steel is hot rolled with final rolling temperature close to the temperature A1. The sheet has fine ferritic-austenitic structure after the rolling and undergoes water quenching to transform austenite into martensite/austenite islands (M/A). Adding Cu into steel composition requires only tuning of process parameters to ensure enough time for Cu to form nanometric precipitates in ferrite grains [4,5].

Dual steel production was designed as rolling below A_3 - at temperatures ranging from A_1 to A_1 + 50 °C. Boron steels was used in the experiment - one alloyed by 1 % Cu and one reference steel without Cu addition. Boron alloying was chosen to slow down proeutectoid ferrite and pearlite formation. The rolling should be performed ideally on the structure of undercooled austenite. This ensures fine ferritic-austenitic structure after the rolling. There was a delay between rolling and quenching at temperatures from 590 °C to 650 °C. The delay was from 11 to 25 sec. long. Any Cu precipitation occurring during that hold increases yield strength of the ferrite and can be measured by tensile testing.

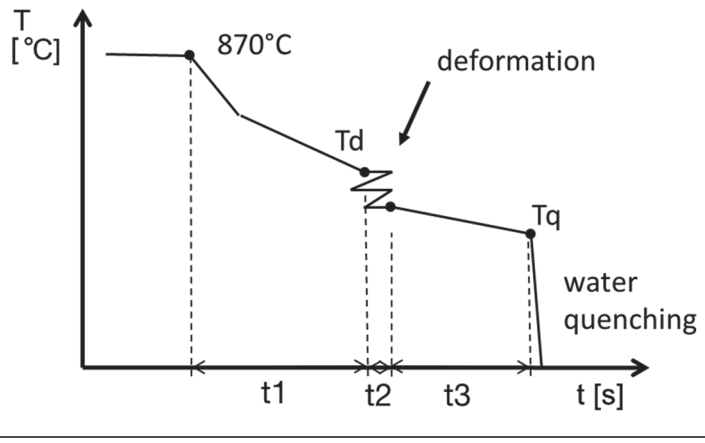
2. EXPERIMENT

There were two experimental steels used for the experiment. Both were casted in induction vacuum furnace under argon atmosphere. Chemical composition of the steels is in **Table 1**. The size of the batches was 500 kg and they were processed by forging and hot rolling into 5 mm thick sheets. Samples for the experiment were cut from the sheets. The samples had length 220 mm, width 75 mm and their longest dimension was in the sheet rolling direction.

Table 1 Chemical composition of the experimental materials in mass %.

Material	C	Mn	Cu	Si	B	Ti	N	Fe
0Cu	0.22	0.98	0.12	0.073	0.0014	0.022	0.0056	bal.
1Cu	0.21	0.98	1.08	0.081	0.0014	0.025	0.0063	bal.

Controlled rolling was performed by experimental rolling mill in duo configuration. The rolls had diameter 550 mm and the rolling speed was 0.25 m/s. Samples were covered by protective coating against decarburization and heated in the atmospheric electric furnace. Cooling of the samples to the rolling temperature (Td) and to the quenching temperature (Tq) was in still air. Final quenching was performed in water bath. Schematic representation of regimes is in **Figure 1** as well as times and temperatures for each regime. Samples labelling is described there too.



Sample	Td	Tq	t3
0Cu_I_L	750	620	21
0Cu_I_S	750	640	11
0Cu_B_L	710	620	19
0Cu_B_S	710	640	12
1Cu_I_L	740	600	21
1Cu_I_S	740	620	13
1Cu_B_L	690	590	25
1Cu_B_S	690	610	16

Figure 1 Scheme of the regimes and parameters for each sample. Sample labelling consists of material sign (0Cu or 1 Cu), temperature designation (I for intercritical deformation, B for deformation below critical temperature) and hold before quenching designation (L for the long, S for the short one). Time t1 varied according to the Td from 12 to 20 sec, t2 was 1 sec.

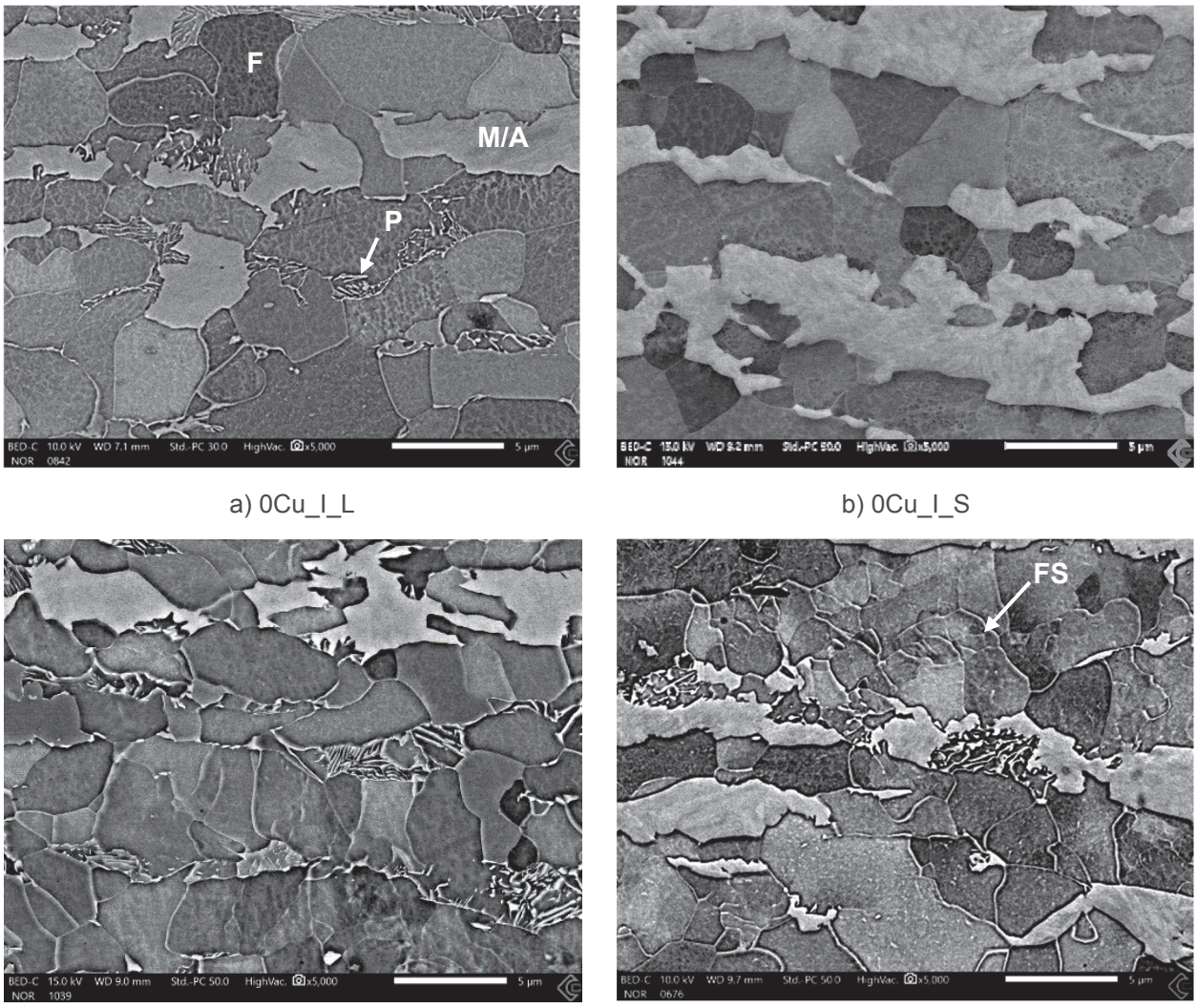
The rolling generally consisted in sample heating to the temperature 870 °C for 20 minutes for complete austenitization. Sample was removed from the furnace and let to cool down to required rolling temperature. The temperature was measured by pyrometer. Two Td were chosen for each material - one in intercritical region (I) and one slightly below A_{c1} temperature (B). The deformation was performed in one reduction from 5 mm sheet into 3.1 mm thickness (i.e. 38 % reduction). The sample was immediately removed from the mill and quenched after certain delay. Two delays were chosen - short one (S) between 11 and 16 sec. and longer one (L) between 21 and 25 sec.

Specimens for the mechanical testing were cut from the samples. Specimens had gauge length 20 mm and width 4 mm. Quasistatic tensile test was performed according to standard EN ISO 6892-1. Proof stress $R_{p0.2}$, tensile strength R_m, total elongation A_5 and reduction of area Z were measured. Three specimens were tested from each sample.

Metallography was performed on longitudinal sections. They were prepared by mechanical grinding and polishing. Final polishing was performed by colloidal silica with average particle size 0.05 μm. Microstructure was revealed by Nital etchant. The microstructure was observed in scanning electron microscope (SEM) JEOL IT-500HR. Backscattered electron signal proved to be the most revealing, particularly to distinguish between ferrite grains and martensite/austenite islands (M/A).

3. RESULTS AND DISCUSSION

The microstructure of all samples was mostly ferritic. The other constituents were M/A islands and pearlite.

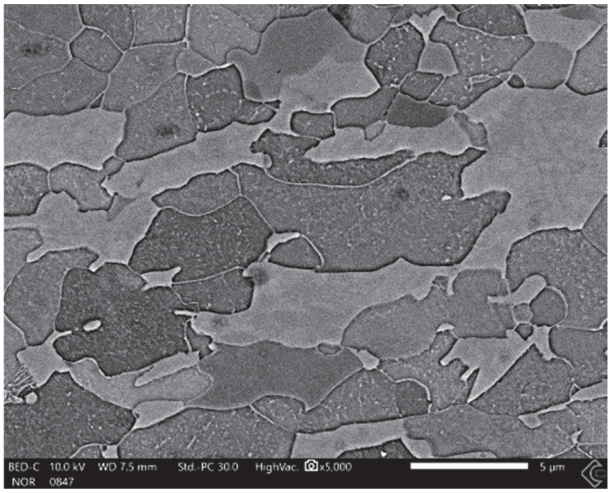


a) 0Cu_I_L b) 0Cu_I_S

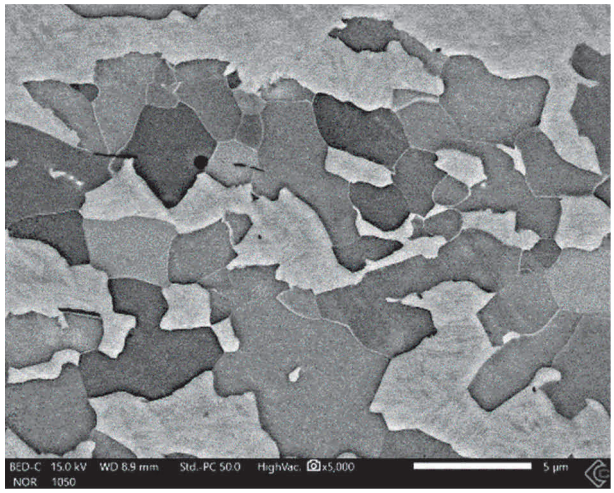
c) 0Cu_B_L d) 0Cu_B_S

Figure 2 Steel without Cu. M/A islands are the bright regions. M/A - martensite/austenite islands; P - pearlite, F-ferrite, FS - ferrite substructure.

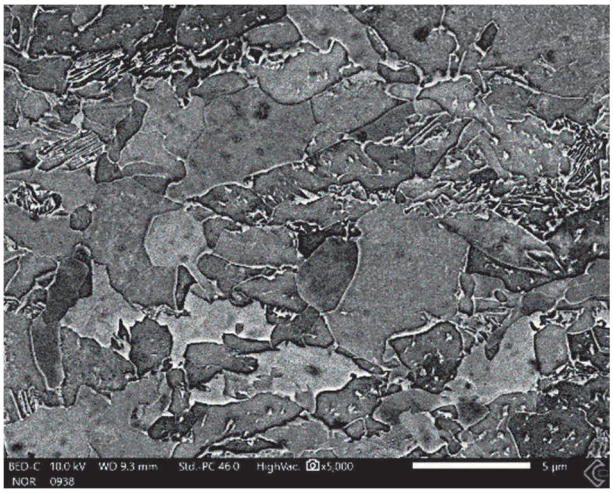
These phases were distributed in bands parallel with rolling direction and partially also randomly among ferritic grains. This is visible in micrographs in **Figures 2 and 3**. Ferritic grains were equiaxed. Ferrite grains in samples “I” (after intercritical deformation) exhibited no visible substructure within ferritic grains. They appeared fully recrystallized. Samples “B” with lower Td exhibited possibly some subgrain boundaries in ferrite grains and the grain boundaries were also not as straight as in case of “I” samples. This indicates that the ferrite recrystallization was not completed at the lower temperature before quenching.



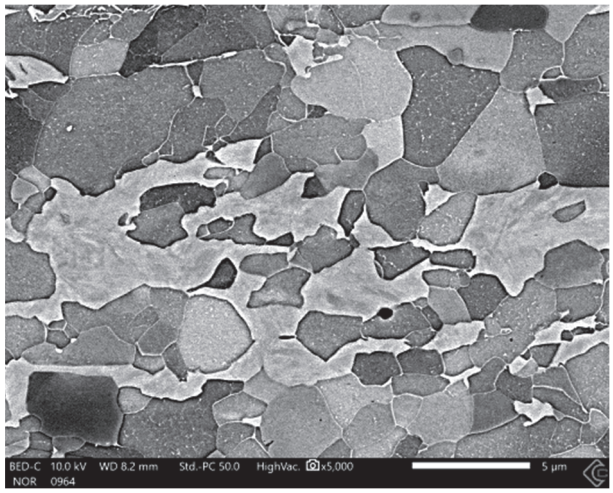
a) 1Cu_I_L



b) 1Cu_I_S



c) 1Cu_B_L



d) 1Cu_B_S

Figure 3 Steel with Cu - microstructures of individual regimes.

Phase content is given in **Table 2**. Micrographs for the quantification were taken in the depth roughly 0.8 mm below the sheet surface. Microstructure close to the surface was different - finer ferrite grains and lower amount of martensite. This was observed to the depth roughly 0.4 mm. There was also decarburized layer 0.1 mm thick on both sheet surfaces. However, this structural heterogeneity was very similar for all sheets. The influence on mechanical properties can be assumed as roughly equal for all samples. Mechanical properties can be attributed to the structures in **Figures 1 and 2** with the remark, the surface layer with different structure alters them approximately the same way for all samples.

Results of the tensile test are in the **Table 2**. Tensile strength showed correlation with the structural composition. Higher amount of M/A islands lead to higher strength, presence of pearlite reduced it. Proof stress $R_{p0.2}$ was not affected largely by the form of hard structure constituent - whether it was M/A or pearlite. Much more important seemed to be the state of ferrite. Fully recrystallized “I” samples exhibited generally lower $R_{p0.2}$ than “B” samples with substructure observed within ferrite grains.

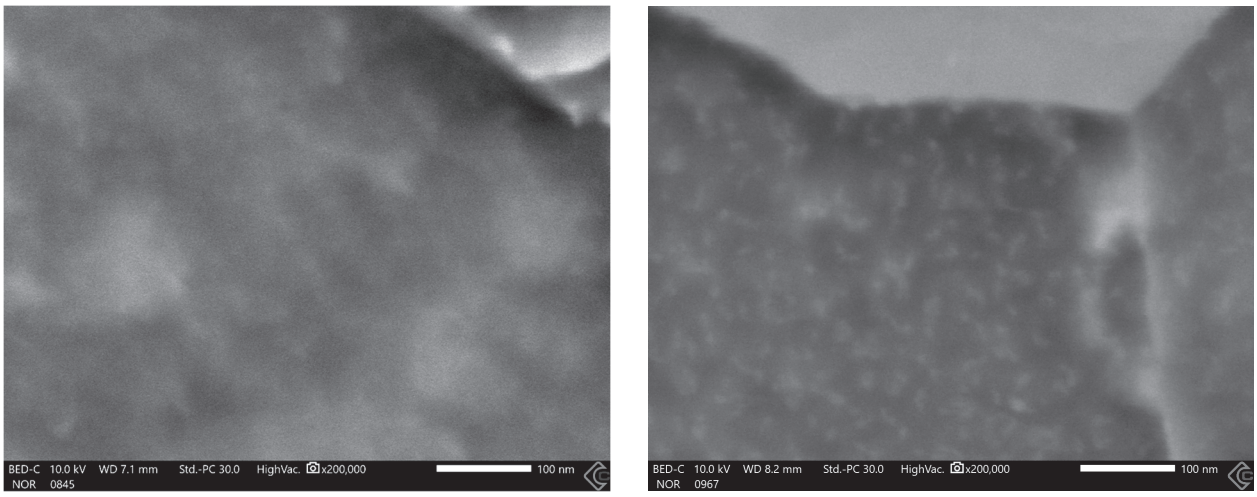
There was also difference between the materials. Longer hold meant decrease of $R_{p0.2}$ for material without copper. This is expected behaviour as the structure had more time for recover and recrystallization during longer hold. Lower amount of lattice defect resulted in lower $R_{p0.2}$. The material alloyed by Cu exhibited

opposite trend. Longer hold increased $R_{p0.2}$ for both deformation temperatures. This increase can be attributed to the precipitation of Cu precipitates within the ferrite.

The best results were obtained for the 1Cu samples deformed below A_1 temperature. $R_{p0.2}$ rose by 40 MPa within 9 sec hold prolongation with no decrease in total elongation and increase of area reduction.

Table 2 Structural composition and mechanical properties of the samples. Proof stress $R_{p0.2}$, tensile strength R_m, total elongation A_5 and reduction of area Z.

Regime	Phases [vol. %]			Mechanical properties			
	Ferrite	M/A	Pearlite	$R_{p0.2}$ [MPa]	R_m [MPa]	A_5 [%]	Z [%]
0Cu_I_L	80	5	15	479	830	12.5	24
0Cu_I_S	70	30	0	487	1022	11.9	18
0Cu_B_L	80	5	15	536	590	24.0	46
0Cu_B_S	75	20	5	598	687	15.6	37
1Cu_I_L	70	30	0	604	932	10.6	19
1Cu_I_S	60	40	0	586	1052	13.2	17
1Cu_B_L	75	5	20	672	843	10.3	27
1Cu_B_S	60	35	5	633	970	10.6	21



a) 0Cu_I_L b) 1Cu_I_L

Figure 4 High-resolution SEM. Scale bars have 100 nm. Distinct bright particles are visible in ferrite grain interior in panel b).

The high-resolution SEM observation was in agreement with tensile test results. **Figure 4** shows the interior of ferritic grains for both experimental materials. The brighter spots - most probably the Cu precipitates - were observed in all samples from 1Cu steel. There were no such distinct features in the 0Cu steel. However, precise characterization could be achieved only by means of transmission electron microscopy.

4. CONCLUSION

Dual and multi-phase structures were prepared by controlled rolling. Proof stress of the samples was mostly determined by state of the ferrite and tensile strength mostly by amount of martensite in the structure. Presence of pearlite instead of martensite significantly reduced tensile strength.

Copper alloying caused higher austenite stability and the steel tended less to form pearlite during the hold after rolling. There was also strengthening observed in copper alloyed steel with longer hold at temperature range 600 - 640 °C. Prolongation of the hold resulted in proof stress increase with simultaneous increase of plasticity and reduction of tensile strength. Copper precipitation had observable influence on properties in time scale 10-25 sec at the holding temperature.

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