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PROCESSING CONDITIONS OF AN ULTRA LOW CARBON/Ti STABILISED STEEL DEVELOPED FOR AUTOMOTIVE APPLICATIONS

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Introduction

Important improvements in the Mexican steelmaking industry have allow the possibility to produce for the first time, ultra low carbon steels (ULC), which are intended for automotive applications [1]. These improvements included (among others) decarburation of molten steels by vacuum degassing. The efficiency of this process, is one of the most important factors for ULC steels, because the carbon content at the end of this process will determine the properties of the final product and the amounts of Ti needed. As has been reported [2,3], vacuum degassed ULC sheet steels achieve very high formability, especially for automotive parts which require good deep drawability. In order to achieve the needed drawability, it has been advised [4–5], that ULC steels should be rolled in the austenite region and the finish discharge temperature, carefully checked, so as not to decrease lower than Ar_3 in order to avoid: (i) abnormal coarse grain structures, (ii) a decrease in the intensity of the (111) texture of sheets, and (iii) deterioration of deep drawability. It is also desirable, that recrystallization of ferrite matrix, of heavily cold rolled ULC sheet steels, results after annealing, in a polycrystalline structure with a strong (111)$<\bar{1}10>$ recrystallization texture, in order to achieve high values of the normal (r) and the average ($\bar{r}$) plastic anisotropy ratio, which is associated with the high formability of ULC steels [6,7].

This work reports advances, on the effect of processing conditions on the mechanical properties of annealed ULC/Ti stabilised steels sheets.

Experimental

The steelmaking route to produce the experimental steel included: (i) the use of 100% high metallization grade sponge iron, which was fed into an electric arc furnace, (ii) vacuum degassing, (iii) furnace ladle treatment, and iv) continuous casting (see Table 1 for chemical composition of slabs). Resulting slabs were re-heated under an argon atmosphere to a rolling start temperature of 1250°C (γ region), finishing the hot rolling operation at 950°C (γ region), reaching at the end of this operation 64% of total

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TABLE 1
Chemical Composition in wt.%

C	N_2	Mn	Si	P	S	Al	Ni	Ti
0.0050	0.0044	0.11	0.030	0.004	0.010	0.071	0.013	0.069

reduction. Coiling of the resulting plate was carried out from 950°C to 730°C (30°C/s) and then cooled to room temperature in air (30°C/hr). Following coiling, plates were cold rolled at room temperature, reaching at the end 84% of total reduction and a sheet thickness of 0.30 mm. Sheet specimens were isothermally annealed under an argon atmosphere at 800°C, from 1 to 1000 seconds. The heating and cooling rate of annealed sheet specimens was ~10°C/s and ~80°C/s, respectively. Specimens were metallographically observed under a 440 Stereoscan scanning electron microscope and a 1200 Jeol transmission electron microscope, and the progression of recrystallization followed by point counting techniques [8].

Flat tensile (ASTM E-8) test on fully annealed specimen was conducted on an Instron 1125 (10 tones) test machine. From the resulted tensile test specimens in the annealed condition, it was calculated the normal and average plastic anisotropy ratio value, at 0°, 45° and 90° with respect to the rolling direction. Texture measurements on hot band, cold band and annealed cold band were carried out on a Siemens D5000 texture goniometer by using cobalt radiation, the pole figures have been measured up to a tilt angle $\chi = 80°$, the steps were $\Delta\chi = 2.5°$ in tilt direction and $\Delta\phi = 4°$ in azimuth direction.

Results and Discussion

In the ULC/Ti stabilised steel was detected the presence of TiN, TiS, $Ti_4C_2S_2$, and TiC precipitates (Fig. 1), which were identified by means of STEM microanalyses. Regarding those precipitates, it can be said that TiN particles form during addition of the FeTi alloy to the liquid steel bath, and after solidification, those precipitates were present at ferrite matrix and at grain boundaries, with sizes between 0.5 to 2.0 μm. After re-heating and hot rolling of slabs in the austenite region, it was detected the presence of TiS and $Ti_4C_2S_2$ precipitates. TiC precipitates were identified after coiling.

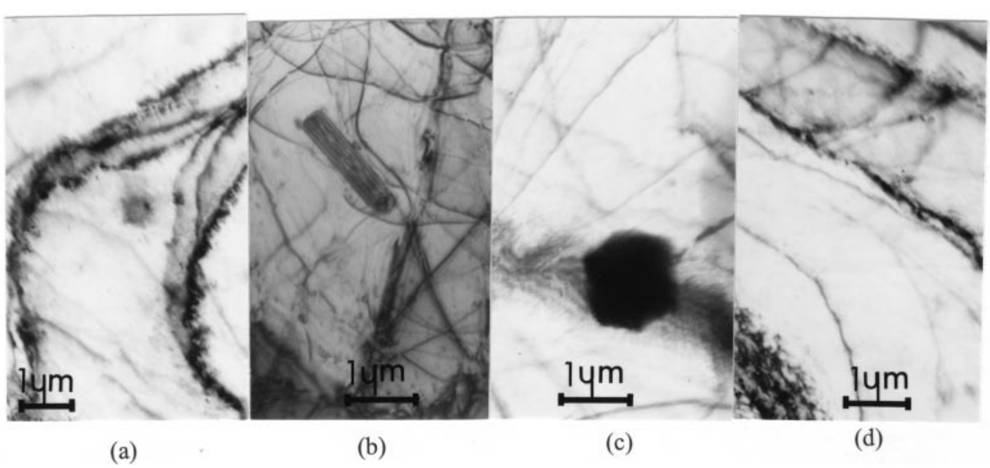
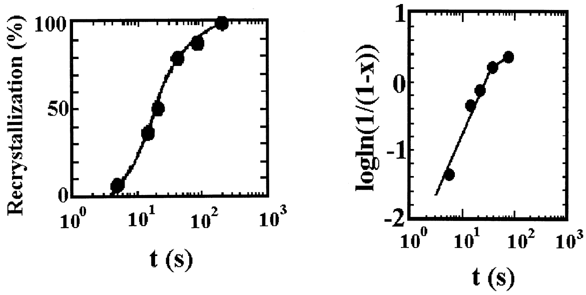


Figure 1. Precipitates detected in ULC/Ti stabilised steel: (a) TiN, (b) TiS, (c) $Ti_4C_2S_2$, (d) TiC.



(a) (b)

Figure 2. (a). Recrystallization curve for the ULC/Ti stabilised steel annealed at 800°C. 2(b). Plot of logln (1/1 − x) vs time.

Regarding TiN precipitates, it has been reported [9] that these precipitates will affect the mechanical properties of resulting annealed sheets. For instance, a fine dispersion of TiN precipitates will modified the growing of recrystallized ferrite grains with {111} texture, being a dispersion of coarse TiN precipitates, the key factor to promote high values of the average plastic anisotropy ratio, $\bar{r}$. For this purpose, it is also advisable to use a high coiling temperature, in order to coarse the ultrafine TiC precipitates that form in ferrite. Sulphides in ULC steels have reported [10] to precipitate as TiS in the higher temperature region of γ and as $Ti_4C_2S_2$ in a lower temperature region of γ (both precipitates were identified in this work), playing this last precipitate an important role to control the solute carbon atoms.

Regarding recrystallization kinetics of cold bands, Fig. 2 (a) shows a recrystallization curve for the cold rolled ULC/Ti stabilised steel at an annealed temperature of 800°C. As can be observed in that figure, annealed specimens reached full recrystallization at about 200 seconds. Fig. 2 (b) shows a plot of logln(1/(1-x)) versus time, from which n values were calculated according to equation of the Kolomogorov-Johnson-Mehl-Avrami type [11]:

$$x = 1 - \exp(-bt^n) \tag{1}$$

$$\mathrm{logln}(1/1 - x) = \mathrm{logb} - \mathrm{nlogt} \tag{2}$$

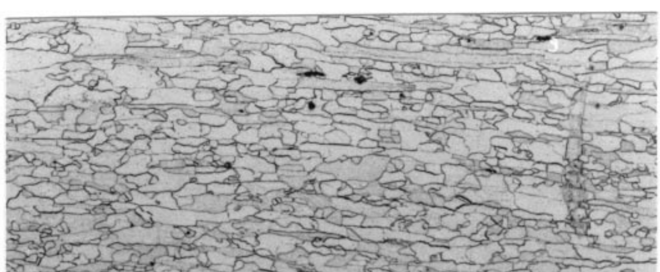
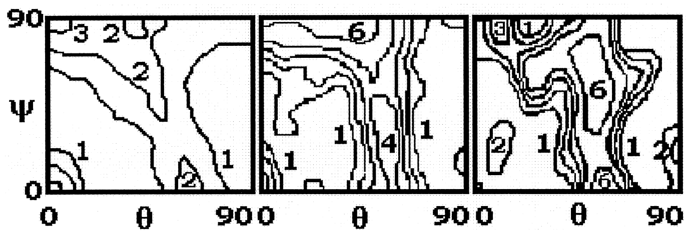


Figure 3. Partially recrystallized ULC steel annealed at 800°C.



(a) (b) (c)

Figure 4. $\varphi_2 = 45°$ section of the ODF's of the (a) hot rolled, (b) cold rolled and (c) annealed bands.

where x is the volume fraction of recrystallized grains, t the recrystallization time in seconds, and b and n are constants related to the geometry and kinetics of nucleation and growth.

As can be observed in Fig. 2 (b), the recrystallization curve of the ULC/Ti stabilised steel showed two well defined stages, each of them characterised by a different n value. For instance, $n = 2.4$ corresponds to the early stages of annealing and $n = 0.47$ corresponds to the later stages of annealing which are characterised by very low values of n.

As was observed in Fig. 2 (a), one of the effects of stabilising the ULC steel with Ti, is in terms of retardation of recrystallization of the cold worked hot band structure. For instance ULC steels with similar compositions [9] showed recrystallization times of about 30 seconds. On the other hand, as was presented above by means of the Kolomogorov-Johnson-Mehl-Avrami analysis, the recrystallization of the ULC stabilised steels proceeds in two stages, and as reported in [12], the low value of n indicates that the morphology of the recrystallized grains are not equiaxed as presented in Fig. 3, but normally, the recrystallization process is so complex to be described with two variables, in particular, the inhomogeneity of the distribution of the store energy and the nucleation events can cause deviations from the Avrami kinetics.

Texture behaviour of hot, cold and annealed bands, is shown in Fig. 4, where it can be seen that cold rolling texture is one of the most important factors affecting the development of {111} recrystallization texture in ULC steels. As is seen in Fig. 4b, the deformed grains with {111} orientation are the most numerous in cold rolled sheets as compared with the {100}<011> texture of hot bands (Fig. 4a). During annealing (Fig. 4c), new grains with {111} orientation nucleate fast and numerously, hence dominate the final texture.

Quantification of the mechanical properties of annealed sheet specimens, in terms of 0.2% yield strength (0.2% Y.S), ultimate tensile strength (UTS) and percentage of elongation (El %), in specimens machined out at 0°, 45° and 90° with respect to the rolling direction (RD), are shown in Table 2. From the results obtained on tensile test specimens in the annealed condition, it was calculated the normal (r)

TABLE 2
Mechanical Properties of Annealed Sheets at 800°C

Mechanical Properties	0° RD	45° RD	90° RD
0.2%Y.S (MPa)	167	156	153
UTS (MPa)	306	295	306
El (%)	56	62	68

TABLE 3
Normal (r) and Average ($\bar{r}$) Plastic Anisotropy Ratio of Annealed Sheet Specimens at 800°C

Plastic anisotropy ratio	0° RD	45° RD	90° RD
r	2.03	2.00	2.08
$\bar{r}$		2.02	

plastic anisotropy ratio at 0°, 45° and 90° with respect to the RD, and from those values, it was calculated the average ($\bar{r}$) plastic anisotropy ratio, as is shown in Table 3.

Conclusions

1. The ULC/Ti stabilised steel responded positively to the hot rolling, coiling, cold rolling an annealing treatment.
2. ULC/Ti stabilised sheet steel in its annealed condition showed excellent properties in terms of elongation and the plastic anisotropy ratio.
3. From the resulting properties, it can be said that the ULC steel can have a wide range of applications, going from commercial quality (%El, 36.9 to 42.2; $\bar{r}$ 1.0 to 1.41) to extra deep drawing quality (%El, 46.1 to 50.5; $\bar{r}$, 1.73 to 2.12), mainly by manipulating both temperature and time of annealing of the resulting sheets, fulfilling the automotive needs.
4. The developed ULC/Ti stabilised steel could compete in terms of mechanical properties, with the modern interstitial free steels, which require a more sophisticated technology.

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