State of the art in the inclusion control within Ca treated Al-killed steels

C. Mapelli, M. Vedani

Nowadays, the clean steel is an important aim for all the steelmakers. The production of the steels with a low content of the non metallic-inclusions allows the users to produce lighter and safer structures as a consequence of the best mechanical behaviour of these materials. But the different non-metallic inclusions affect the mechanical properties of the steels by different ways. So, a good design of the materials is founded on the precise knowledge about the influence of the non-metallic inclusions in several situations as a function of the compositions and shapes of these compounds. The melting and the refining of the metal bath is accurately controlled in order to eliminate the possible sources of the non-metallic inclusions, decreasing the gas solved in the liquid bath by foaming slag, vacuum treatment, deoxidizers etc.. However, the use of these last elements can also produce dangerous compounds that have to be modified by Ca treatment or other new methods devoted to change the shape and chemical compositions of the inclusions in order to avoid their detrimental effects on the material properties or the interruption of the continuous casting operation for the occurrence of the nozzle clogging. The oxygen content at the beginning of the steel affination and its control are certainly one of the most important factors to control the genesis and modification of the non-metallic inclusions.

Parole chiave: acciaio, colata continua, microscopia elettronica

INTRODUCTION

The mechanical effects of the non-metallic inclusions on the mechanical properties of the steel and the mechanism of interaction between the metal matrix and the non-metallic compounds are one the most important aspects affecting the steel reliability. The content of non-metallic inclusions, their chemical composition and the acceptable amount of these compounds is defined on the basis of the final use. So the design of the process is linked to the microstructural and mechanical features of the final product, and the understanding of the thermodynamic and chemical principles ruling the genesis and the growth of non-metallic inclusions.

In the last decades the use of continuous casting system to improve the production rate has known a large diffusion, but some efforts were and are performed to increase the regularity of this process. One of the most detrimental cause of the decrease in the steel plant efficiency is the nozzle clogging that can imply a too early stop of the casting operations. This phenomenon is caused by the accumulation of the non-metallic inclusions within the nozzle and the most detrimental inclusions are mainly due to the compounds formed during the deoxidation procedure.

The addition of the calcium during the refining period is the most diffused procedure to avoid the nozzle clogging by the formation of calcium-aluminates with low melting point, which cannot produce this undesired phenomenon. Modification of inclusions by Ca in Al-killed steels was extensively studied in the last decades. At the usual steel refining temperatures, the most important metallurgical effects are associated with modifications of solid alumina inclusions to liquid calcium aluminates in order to alleviate nozzle clogging concerns during continuous casting. MnS inclusions are concurrently modified to (Ca,Mn)S compounds that can exist as a single phase or can surround calcium aluminate aggregates.

However, the mechanical problems can rise up for the mechanical characteristic of the non-metallic inclusions, that can be too weak or can generate some detrimental effects related to their interactions with the metal matrix, not only in the final applications but even during the forming operations based on the plastic deformation. In the present study the system involving Fe-Mn-Si-Al-Ca-O-S by a thermodynamic model to define the best condition for the steel castability and a technological alarm the operators can keep under control to reach the aimed inclusion composition is here proposed. The thermodynamics treatment is based on the settlement aquired by the DALM model [4].

The formation of even more complex non-metallic inclusions with the presence of Si to form calcium aluminium silicate are not treated because they are the object of an on going study.

C. Mapelli, M. Vedani
Dipartimento di Meccanica del Politecnico di Milano

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MECHANICAL EFFECT OF NON-METALLIC INCLUSIONS

The mechanical consequences of the presence of the non-metallic inclusions is well described by the Gladman's approach [1] which regards the non.metallic inclusions as discontinuity in the metal matrix, which can appear like holes. The mechanical effect of the non-metallic inclusions can be summarized in three main modalities:

- holes;
- pressurized holes;
- self-pressurised holes.

The first case is treated by the basic analysis proposed by Plateau and Gurland [2], whose model is based on the ductile fracture produced by the genesis of the voids and their growth at the non-metallic inclusions after the fracture of the non-metallic inclusions themselves or by the decohesion between the non-metallic inclusions and the surrounding metal matrix. When a metal containing the non-metallic inclusions is under the action of a state of stress, it can reach a critical deformation leading to the void generation in the metal matrix starting from the non-metallic inclusions.

Some relations were developed for the decohesion and the inclusion fracture and these efforts permit to underline the most important factors of influence on this fracture and void generation process. The most important aspect of these relations is that the size of the inclusions has not any influence on the fracture process, unlike the volume content of the inclusions and their shape which can have a significantly different behaviour as function of the interaction between the principal stress direction and the ratio between the principal axis of the non-metallic inclusions. This last statement is the source of the inclusion induced anisotropy within the steel. It is worth noting that even the inclusions can produce a brittle mechanical behaviour, the basic microstructural weakening process is a ductile one.

The inclusions work as pressurised holes when the gas in solution within the metal shows a lower solubility in the steel than that in the non-metallic inclusions as the decrease of the temperature takes place. A significant example of this situation is represented by the hydrogen migration from the metal to the sulphide inclusions, as a consequence of the decreasing of the solubility product of the gas element as the temperature becomes lower and lower. The migration of the gas produces an increase of the gas pressure within the non-metallic inclusions and the related stress between the inclusion and the metal matrix. In certain situations the role of the non-metallic sulphides as tank of hydrogen can be searched when some hydrogen brittleness is feared and the inclusions can operate as sucking unity of the gas element solved in the steel matrix.

Finally, the non-metallic inclusions could behave as self-pressurised holes. This behaviour is due to the different thermal dilation coefficient between the metal matrix and the non-metallic inclusion that are ceramic or glassy-ceramic compounds. There are three possibilities:

- the metal matrix and the inclusion have the same thermal dilation coefficient: no induced stress;
- the metal matrix has a greater thermal dilation coefficient than the one of inclusion: tensile state of stress on the metal matrix;
- the metal matrix has a lower thermal dilation coefficient than the one of inclusion: compressive state of stress on the metal matrix.

The different non-metallic inclusions have different thermal coefficients [3]. Because of this third mechanism is due to the thermal expansion phenomenon, the size of the non-metallic inclusions has an important influence on the magnitude of the state of stress that can be generated and the residual stress remaining in the matrix can affect important mechanical features such as the fatigue limit of the steel that is a formerly stressed structure.

These evaluations point out that the inclusion engineering is a fundamental task for the production of the steel showing the required mechanical characteristics. So, the inclusion precipitation has to be ruled in order to produce the precipitation of non-metallic inclusions featured by a composition that allows to improve their mechanical behaviour to avoid decohesion, detrimental shape after deformation process and the rising of self-tensioned state due to the different thermal dilation of the non-metallic inclusions. The modifications of the non-metallic inclusions implemented in the steel production for continuous casting aimed at two main achievements:

- non-metallic inclusions with low melting point;
- mechanical behaviour of the non-metallic inclusions suitable for the final application and structural reliability.

By a correct Ca treatment it is possible to achieve these two aims. However, the present study is devoted mainly to the good castability of the steels, while the inclusions with the best mechanical behaviour require also the presence of a correct quantity of silica within the non-metallic inclusions. The model about this last task is object of an on going study. On the other hand, in this study will be briefly treated the problem of the sulphide crown on the oxide inclusions, because even this phenomenon can contribute to improve the mechanical performance of the metal matrix-inclusions system.

MATERIAL AND EXPERIMENTAL PROCEDURE

The experimental tasks have been developed on steels with different composition (table I). Particularly, the sulphur content varies significantly. The material investigated in the present study was a re-sulphurised C-Mn Al-killed steel and two not re-sulphurised steels whose composition differs from the first one mainly for the sulphur content. The steel features a ferritic-pearlitic microstructure.

The validation of the proposed model is implemented on these different cases to allow the validation of the thermodynamics model in a wide range of composition to evaluate the possibility to extend the proposed approach to a large chemical composition field. Particularly the different S content can become a very important aspect, because of the great interaction of this element with the calcium and its repartition between CaS and CaO.

The steel is produced by scrap melting in a 100t EBT (eccentric bottom tapping) electric arc furnace. Secondary steelmaking is carried out in a 100t ladle furnace. In particular, after the initial deoxidation, the standard ladle furnace treatment consists in desulphurisation with CaO and CaSi alloy and concurrent argon bubbling through nozzles positioned on the ladle bottom that follows over-heating and temperature setting. The steel 1 is resulphurised by FeS wire addition and, subsequently, Ca treated by CaSi injection. Synthetic slag had the following average composition: 51% CaO, 26% Al_2O_3, 12%SiO_2, 9% MgO, other constituents in minor amounts being FeO_n, MnO, Cr_2O_3, P_2O_5, TiO_2. After refining, the steel is continuously cast in a curved four-strand machine to produce billets having diameters ranging from 145 to 280 mm. Fifteen heats of each type of steel were monitored by sampling "sucked" specimens (disks having a diameter of 25 mm and a thickness of 10 mm) from the ladle at the end of the refining practice and from the tundish during continuous casting. The small volume of metal taken for each sample guaranteed a rapid solidification of the specimen on withdrawal from liquid metal thus preserving the actual metal-inclusion state of ladle environment. The samples were analysed using an inductively coupled argon plasma spectrometer to obtain the chemical composition. Single use electrochemical oxygen cells (Celox) were used to evaluate oxygen activity in the ladle. Thermocouple allowed measuring of the actual steel temperature at the time of sample collection. The obtained samples were sectioned and

%wt	C	Mn	Si	P	S	Al	Ca
Steel 1	0,17	1,29	0,5	0,015	0,02	0,02	13
Steel 2	0,15	1,25	0,5	0,01	0,005	0,025	15
Steel 3	0,15	1,25	0,4	0,015	0,001	0,025	14

TableI The average composition of the analysed steels

Tab. I Composizione media degli acciai studiati

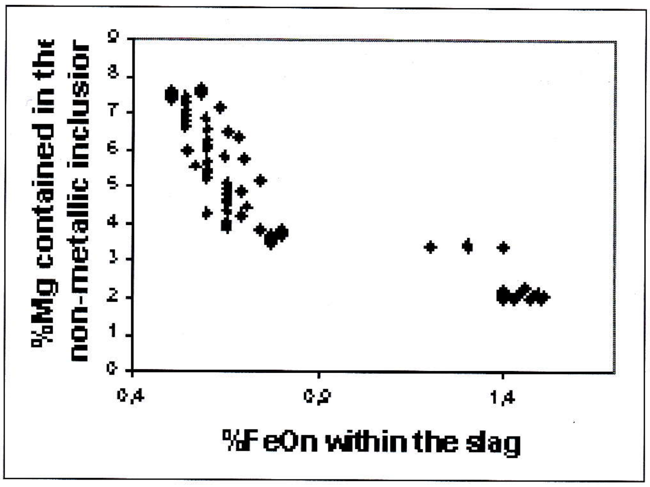

Fig.1 Mg in the non-metallic inclusions vs. initial %FeO$_n$ in slag

Fig.1 Contenuto di Mg nelle inclusioni non metalliche in relazione a %FeO$_n$ contenuto in scoria

prepared for metallographic observation by grinding and polishing.

After preliminary optical microscope analyses for detection of inclusion distribution, the samples were observed in a scanning electron microscope (SEM) equipped with an X-ray energy dispersive spectrometer (EDS) system. Inclusions larger than 1 mm were analysed quantitatively as a matter of routine, whereas qualitative spectra were occasionally taken from smaller inclusions for their identification. A sample of the slag formed at the beginning of the refining ladle treatment has been kept and analysed by a X-ray analyses to measure the concentration of the FeO$_n$, whose content is then related to the total oxygen measured in solution within the metal bath at the beginning of each treatment in ladle furnace.

The above experimental method allowed to gather data on the chemical composition of the steel at the end of the steelmaking practice and, correspondingly, to investigate the inclusion morphology and composition detected in the collected samples. The temperature of this final stage is within the range 1833±5K. Data on chemical analysis and temperature of the steel were used as input for a model validated in some previous tasks [4]. The theoretical prediction on type and amount of inclusions formed in the steel was then compared with the experimental results.

EXPERIMENTAL RESULTS

The quantitative chemical analysis by EDS was performed on 500 inclusions from the sampled material.

The inclusions of the steel 1 is featured by small CaS or (Ca, Mn, Fe)S inclusions containing a significant fraction of Fe and Mn were detected in all the samples. As expected, a large amount of MnS inclusions was found. These inclusions were generally dispersed in the steel as isolated particles having micrometer or sub-micrometer size. The simultaneous presence of the two strong deoxidisers as Al and Ca led to the precipitation of $YCaO.XAl_2O_3$, then indicated as CA_x. However, the high activities of S and Ca in this phase also led to the partial deposition and precipitation of CaS on CA_x. Duplex inclusions having a CA_x nucleus and a surface region enriched CaS but also of MnS were detected. The MnS inclusions show also some traces of Ca, probably combined in the form of CaS that can contribute to make globular the MnS compounds. In addition are detected several single CaS inclusions. Within the nucleus it is pointed out also the presence of some traces of Mg with content that varies from 2% to 8% in the different samples. It is possible to suppose that the Mg is combined in the form of $MgO.Al_2O_3$ spinel. The content of Mg was found to increase in the heats where the beginning %FeOn in the slag and the corresponding oxygen activity of the metal bath are lowest (fig.1).

The heats characterized by the other two compositions show similar results except for the presence of nearly pure CaS inclusions and the decreased content in the crown covering the CAx ones. Moreover, the presence of a few examples of MnS was found in two heats of the steel 2, while in the steel 3 an absolute absence of these inclusions has been pointed out. The Mg presence within the nucleus of CAx is verified also in the sample belonging to the steel 2 and steel 3 and the content range of this presence is the same for the steel 1. The inverse trends between the initial %FeOn and the oxygen activity of the slag with the Mg content is confirmed also in the samples of the steel 1 and of the steel 2.

DISCUSSION

The high amount of sulphides in the steel one is certainly due to the high sulphur content of this steel that is re-sulphurised grade. The presence of the CaS nearly pure inclusions is probably the consequence of the introduction of sulphur in high temperature range, whereas the high temperature (>1863K) favours the partition of sulphur in the CaS form than in the CaO one as indicated by the equilibrium constant for the two reaction:

$\underline{Ca} + \underline{O} \leftrightarrow (CaO)$(1)

$$\log K_{CaO} = \log a_{CaO} / (h_{Ca} \cdot h_O)^{(2)} = 25655/T - 7{,}65 \qquad [5]$$

$$\underline{Ca} + \underline{S} \leftrightarrow (CaS)$$

$$\log K_{CaS} = \log a_{CaS} / (h_{Ca} \cdot h_S) = 19980/T - 5{,}9 \qquad [6]$$

with the global reaction that can be written as:

$$(CaO) + \underline{S} \leftrightarrow (CaS) + \underline{O}$$

whose equilibrium constant

$$K' = (a_{CaS} \cdot h_O) / (a_{CaO} \cdot h_S) = K_{CaS} / K_{CaO}$$

where the last ratio increases as the temperature goes up.

The MnS precipitation takes place in a steel with such a composition as a consequence of the trend of the equilibrium constant for the reaction:

$$\underline{Mn} + \underline{S} \leftrightarrow (MnS)$$

$$\log K_{MnS} = \log a_{MnS} / (h_{Mn} \cdot h_S) = 9281/T - 6{,}43 \qquad [7]$$

that allows a significant precipitation of the MnS after the S addition and with the Mn content typical of the steel 1.

The sulphide deposition that builds a circular crown on CA_x happens and is favoured because the precipitation of the sulphides takes place after the formation of the complex oxide inclusions. Actually the thermodynamic rule that controls the sulphide precipitation within the liquid metal bulk is the

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(1) within the bracket are indicated the phases different from the metal bath.

(2) the underscored compound and element are solved in the metal bath and constitute with that a unique phase
a_x = raoultian activity of the x element
h_x = henrian activity of the x element

balance between the free volume energy and the free surface energy to form the non-metallic aggregates within the liquid metal:
where

$$\Delta G = -\frac{4}{3}\pi r^3 \frac{L\Delta T}{T_m} + 4\pi r^2 \gamma_{s\to l}$$

ΔG is the free energy variation for the formation of a solid phase in a liquid metal bulk, that is supposed to be a sphere;
r is the radius of the sphere;
L is the latent heat per unit volume developed by the solidification;
ΔT is the super-cooling of the precipitating compound;
T_m is the theoretical melting point of the precipitating compound;
$\gamma_{s\to l}$ interfacial energy between the solidified compound and the metal bulk.

Because the ΔG has to be negative, the limit condition for the precipitation process to occur is the balancing by volume term of the positive term given by the interaction between the solid surface and the liquid metal bath due to the action of the $\gamma_{s\to l}$ term.

Generally, a large under-cooling (ΔT) can be needed to produce the non-metallic compound precipitation to balance the effect on ΔG of the interfacial energy that is predominant for low value of r. It is well known that a value of r exists that marks a transition from the positive values of ΔG to the negative ones and it is known as the critical radius for precipitation of a solid phase [8]. On the other hand, the former presence of the oxides within the metal bulk can represent a preferential nucleation site for the formation of the sulphides that cover a formerly precipitated compound, whose radius is greater or a bit smaller than the critical radius for the precipitation. So that if the oxides precipitate before the addition of S, the MnS compound and other sulphides (i.e. CaS) will find the oxides inclusions as preferential nucleating site of deposition for themselves. This phenomenon is an advantageous one, because the deposited sulphides can compensate the contrary thermal dilation process of the oxide compounds that become coated avoiding the development of the residual tensions due to the different dilation behaviour between the inclusion and the metal matrix.

The second important aspect is the presence of a certain content of Mg in the nucleus of the CA_x inclusions. This aspect can be explained by the dissolution of MgO.CaO refractory material or more probably by the dissolution of the MgO residuals of the slag of the previous heats that solidified on the refractory walls and that melt and dissolve again in the liquid metal bath when the ladle is filled again. Certainly, the final Mg content of the bath is controlled by the kinematics parameters that rule the rate of this phenomenon, but also a thermodynamics hypothesis about the relation between Mg content and the beginning $\%FeO_n$ slag content or the related initial oxygen activity can be proposed.

The $\%FeO_n$ can be related to the oxygen activity of the steel bath:

$$\log K'' = \log (a_{FeOn})/a_O = 6320/T - 2{,}765 \qquad [9]$$

and

$$\log K''' = \log a_O/(p_{O2})^{0{,}5} = 5832/T + 0{,}356 \qquad [10]$$

so that it is possible to compute the oxygen potential of the steel bath starting from the $\%FeO_n$ and from its activity in the slag, which is computed by the model proposed by Ohta and Suito [11].

This approach is validated from the data of the measured oxygen activity at the beginning of the ladle furnace treatment that seem to fit well with the computed ones on the basis of the %FeOn contained in the slag (fig.2).

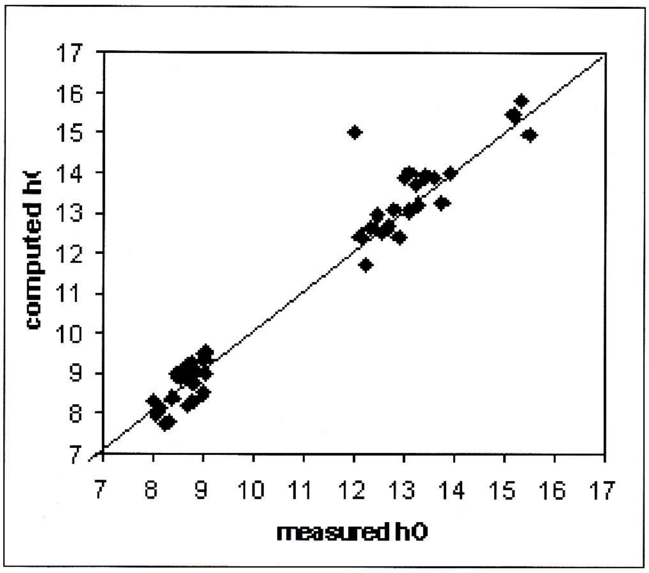

Fig.2 Comparison between the measured oxygen henrian activity and the computed ones (10^{-4}).

Fig.2 Confronto fra le attività henriane calcolate e quelle misurate per l'ossigeno (10^{-4}).

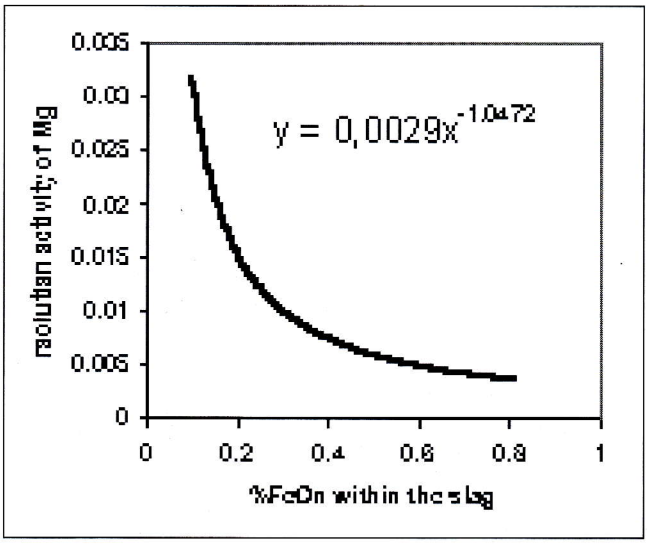

Fig.3 Computed activity of Mg in steel bath vs. $\%FeO_n$ in the slag at 1873K

Fig.3 attività calcolata del Mg vs. $\%FeO_n$ in scoria (1873K)

The free energy of formation of MgO is computed on the basis of the data provided by Knacke et al. [11] (table II).

The oxygen potential of this reaction within the steel bath can be determined under the hypothesis that the MgO residual are pure and so it can be stated that $a_{MgO}=1$ and the Mg content in the steel bath is nearly null (i.e. $\%Mg=1\cdot10^{-4}$). Before the deoxidation, the activity of the oxygen within the metal bath depended on its equilibrium with the slag. As it has just been said, a reliable index of the amount of the oxygen provided by the slag to the bath before the deoxidation is the activity of FeO_n in the slag. So that, if this last content is known, it will be possible to compute the oxygen potential of the bath and comparing that with the standard free energy formation of the MgO. The oxygen potential of MgO formation reaction (in the hypothesis that $a_{MgO}=1$ and $a_{Mg}=2\cdot10^{-6}$) is positive above 1725K so that the dissolution of the MgO in Mg and O can occur. The dissolution of Mg

TableII. Standard free energy formation for the reaction Mg + O →MgO

Tab.II Energie di formazione in condizioni standard (a_{Mg} = 1, a_{MgO} = 1) per la reazione

Mg + O → MgO						
T (K)	1323	1433	1543	1653	1763	1873
ΔG (J/mol)	-495036	-458410	-421785	-385159	-348534	-311908

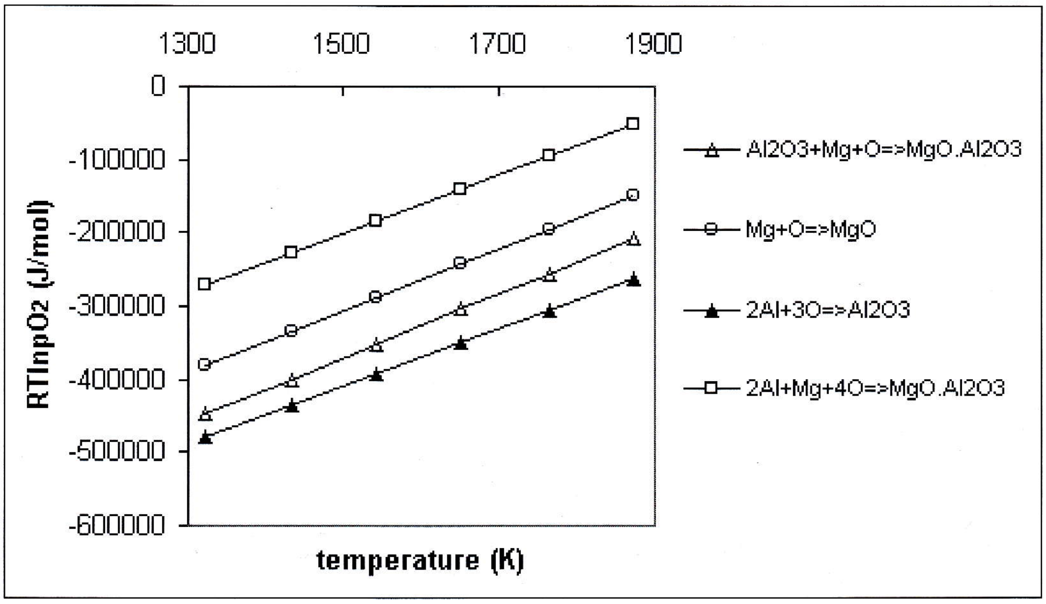

Fig.4 Oxygen potential for different reaction of oxidation (h_{Al}=0,022%, a_{Mg}=0,005, activity of all the oxides is equal to unity because they are considered as pure phases)

Fig.4 Potenziale di ossigeno per diverse reazioni di ossidazione di interesse (h_{Al}=0,022%, a_{Mg}=0,005, l'attività degli ossidi è considerata unitaria)

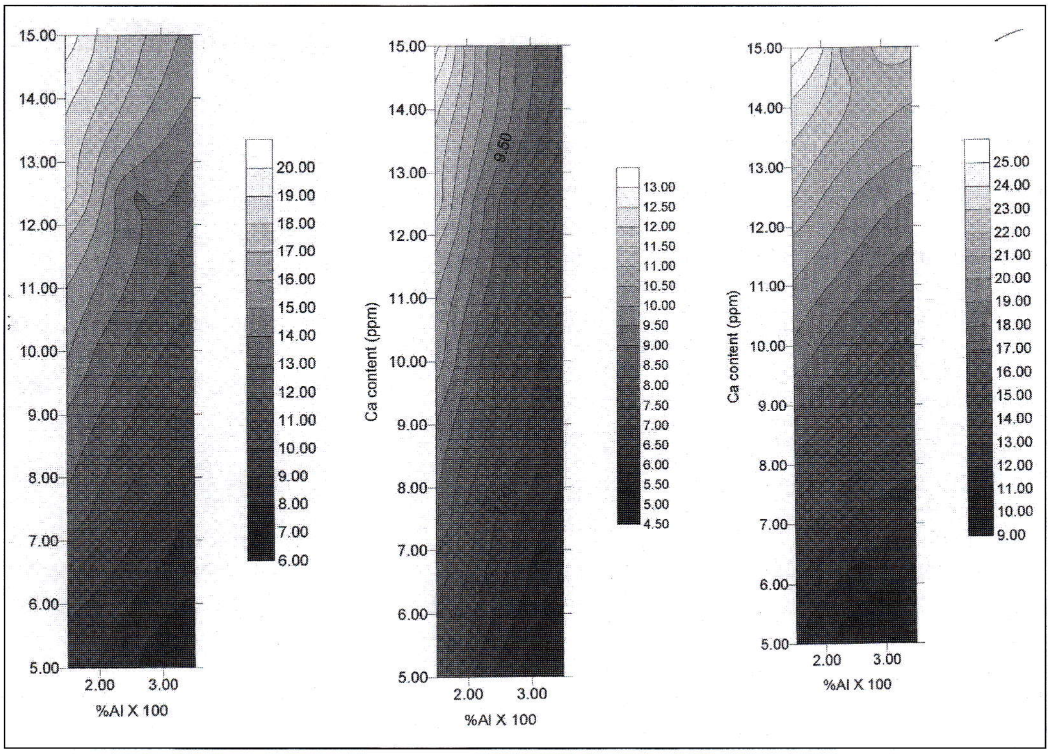

Fig.5 Three different map for the optimal %FeO_n (x100) content for three different S content (A_%S=0,005%, B_%S=0,02%, C_%S=0,001%) and the chemical composition and slag typical of the three studied steels.

Fig.5 Tre differenti mappe per stabilire il contenuto ottimale di FeO_n (x100) per tre differenti contenuti di S (A_%S=0,005%, B_%S=0,02%, C_%S=0,001%) e composizione chimica della scoria e dell'acciaio tipica dei tre acciai studiati.

from the refractory wall to the metal bath is thermodynamically possible till the oxygen potential for the Mg oxidation reaction is equal to that of the metal bath. The Mg can dissolve in the bath where it can combine itself with Al_2O_3 to form the spinel precipitates. The lower %FeOn contents requires higher activity of Mg in the bath and the related higher Mg content (fig.3).

Maybe, the MgO residuals of the slag are the main causes of the Mg dissolution in the bath because of their consistence that can promote fast dissolution process. Actually, the formation of the $MgO.Al_2O_3$ spinel seems to take place in first stage of the ladle furnace treatment, because their presence is within the nucleus of the other oxides and it is possible that this spinel compounds can play the role of a preferential site of nucleation for the successive formed oxide compounds. The spinel has the possibility to form during the deoxidation with Al after the dissolution of Mg in the bath, because on the basis of the thermodynamic data the standard free energy is computed for the reaction:

$(Al_2O_3) + \underline{Mg} + \underline{O} \rightarrow (Al_2O_3.MgO)$

and it is also computed after considering a_{Mg}=0,005 (fig.4), without variation for the activity of the other compound. The reaction probably takes place according to the former reaction, because it is not possible to suppose that the path of the reaction is:

$$2\underline{Al} + \underline{Mg} + 4\underline{O} \rightarrow (Al_2O_3.MgO),$$

because this last reaction requires a too much great value of the Al activity that is never reached in the steel studied even after the Al addition, because with a composition like that of the steel and with a Al=0,02% the model previously validated [4] yields an average value h_{Al}=0,022% and the related $a_{Al}=1,56 \cdot 10^{-5}$. Provided that $MgO.Al_2O_3$ has really formed, the spinel precipitates during the formation of pure Al_2O_3. Moreover, the starting oxygen activity and the related %FeOn content at the beginning of the ladle-furnace treatment are computed so to assure the formation of the correct ratio of CaO and Al_2O_3 to reach the desired and low melting point form $12CaO.7Al_2O_3$. The computation is based on the previous mentioned model (fig. 5) and has been performed with reference to a bath at 1833K. The computations are performed for three different sulphur contents, like the steels object of this study.

It is worth noting that the initial required oxygen that has to be provided by the slag to produce the correct ratio of the CaO and Al_2O_3 is the lower the higher is the sulphur content, because a part of Ca is combined under the formation of CaS and the quantity of CaO formed is less, so that the Al_2O_3 amount to be formed is smaller. Thus, the free oxygen content to be combined can be enough even in smaller quantity, too.

CONCLUSION

On the basis of the analyses developed on three different types of steel it is possible to underline a brief state of the art to assure the correct engineering of the non-metallic inclusions in Ca-treated Al-killed steels.

- The sulphide crown at the oxides surface can be obtained more easily if the S content is increased after the complete formation of the oxide inclusions. It is important to take the care to increase the sulphur content at not too high temperature (>1863K), because this can produce a great formation of pure CaS and the deposition of sulphur on the CA_x can modify the ratio between the two forming compounds altering also a good and desired CA_x.
- The formation of $MgO.Al_2O_3$ spinel can be favoured by low oxygen activity (and the related low oxygen potential), because this stimulated the dissolution in the bath of Mg belonging to the residuals of the slag or to the refractory walls.
- The formation of $MgO.Al_2O_3$ spinel within the metal bulk follows the Mg dissolution and takes place during the Al deoxidation when the precipitation of pure Al_2O_3 happens. The chemical and the most reliable chemical reaction that happens is:

  $$(Al_2O_3) + \underline{Mg} + \underline{O} \rightarrow (Al_2O_3.MgO)$$

  because it is the only thermodynamically possible reaction for the formation of $Al_2O_3.MgO$ spinel.
- The $\%FeO_n$ content of the slag can be taken into account as a reliable technological alarm for the correct development of the deoxidation process and Ca treatment of the steel. The correct $\%FeO_n$ content has to be evaluated on the basis of the slag components that can modify the activity of FeO_n. Moreover, it is important the evaluation of the aimed sulphur content of the steel because this parameter can alter the partition of Ca between CaO and CaS.

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ABSTRACT

ABSTRACT

Oggi l'acciaio pulito è un obiettivo importante per i produttori di acciaio. La produzione di acciai a basso contenuto di inclusioni non metalliche permette agli utilizzatori di produrre strutture più leggere e più sicure, in conseguenza del miglior comportamento meccanico di questo materiale. Ma le diverse inclusioni non metalliche influenzano le proprietà meccaniche degli acciai in modi diversi. Così una buona progettazione del materiale si fonda sulla precisa conoscenza riguardante l'influenza delle inclusioni non metalliche in diverse situazioni ed è funzione delle composizioni e delle forme di questi composti. La fusione e l'affinazione del bagno metallico sono controllate accuratamente per eliminare le possibili fonti di inclusioni non metalliche, diminuendo il gas disciolto nel bagno liquido attraverso scorie schiumose, trattamento sottovuoto e disossidanti ecc. Tuttavia, l'uso di questi ultimi elementi può anche produrre composti pericolosi che devono essere modificati attraverso trattamenti al Ca o attraverso altri nuovi metodi atti a cambiare la forma e la composizione chimica delle inclusioni al fine di evitare i loro effetti negativi sulle proprietà del materiale o l'interruzione delle operazioni di colata continua a causa di ostruzioni. Il contenuto di ossigeno iniziale ed il suo controllo sono uno dei fattori più importanti per controllare la genesi e la modificazione delle inclusioni non metalliche.