1. Introduction

Advantages of the semi-solid casting process have generated significant interest in recent years among casting manufactures. When compared with conventional casting or solid-state forming processes, the semi-solid casting process has been recognized as a technique offering several advantages, such as the prevention of macrosegregation and porosity, and relatively low forming temperatures.\(^1\)–\(^3\) The typical non-dendritic structure constituted of globular solid phase surrounded by the liquid is at the origin of these interesting properties, allowing the semi-solid alloys to be handled as a solid when at rest and attaining fluid like properties during the forming operation. Therefore, the study on the formation of the globular or non-dendritic structures becomes an important subject in this innovative metal technology.

Recently, extensive efforts have been made in order to investigate the mechanism of evolution of globular structure. Some experimental observations such as fragmentation of large crystals, the aggregation, coalescence and abrasion of the fragmented crystals are considered to be responsible for such a spherical structure formed in the processes such as mechanical or electromagnetic stirring, intensive ultrasonic treatment during solidification or partial remelting of deformed alloys.\(^4\)–\(^7\) However, for a recently developed new method—directly obtaining the globular structure from the molten alloys by controlling pouring temperature and cooling rate, there is a shortage of explanation and demonstration about the mechanism of globular structure formation during this simple process without any mechanical actions on the alloy slurry. Obviously, this new non-dendrite structure fabrication process without any additional operation and special equipment has much more significant practical advantages than the previous techniques. It is therefore necessary to clarify and quantitatively understand the evolution of dendritic and globular structures in solidification of alloys.

The recent advances of numerical modeling techniques have made it possible to analyze transport phenomena, including heat, mass and fluid flow in the mushy zone during solidification, leading to the development of various kinds of deterministic and stochastic models for the prediction of microstructure evolution in solidification of metals. Compared with the deterministic models, the stochastic models, including Monte Carlo models\(^8\) and Cellular Automaton models,\(^9\) have the advantages of depicting the microstructure features graphically. Over the last ten years, several studies based on the cellular automaton models have been reported on the evolution of solidification grain structures in various solidification processes.\(^10\)–\(^13\) Recently, a modified cellular automaton model, which includes the curvature effect and the solute redistribution both in liquid and solid during solidification, was developed to predict the
formation of dendritic structures in solidification of alloys.\textsuperscript{[14]}

The purpose of the present study is to study the morphology evolution of the primary phase, i.e., dendritic, intermediate rosette-like and globular, in solidification of Al–7\%Si alloy using a modified cellular automaton (MCA) model. The effects of pouring temperature, cooling rate and inoculation on the morphology of the primary phase were examined. The simulated results were compared with those obtained experimentally.

2. Experimental Procedure

An Al–7.0\%Si alloy was prepared using commercial pure aluminum (99.9\%) and metallic silicon in an electrical resistance furnace. The inoculation was achieved by the addition of Al–5Ti–B master alloy to the level of 0.01 mass\% Ti. The melt temperature was initially kept at 700°C for a specified period, and then slowly cooled down to various temperatures in a crucible. Subsequently, the melt was poured into a preheated steel mold having a thermally insulated bottom placed in a temperature holding furnace in order to cool the specimen with various cooling rates. The experimental apparatus is shown in Fig. 1. A thermocouple was positioned in the center of the mold cavity to measure the cooling curves of the slurry. The temperature-time curve of the slurry was recorded using a data-acquisition unit. The slurry specimen was quenched into water when it was cooled down to a specified temperature. The specimens for metallographic observation were cut from the area near to the thermocouple, polished and then etched with Tucker solution.

3. Mathematical Formulation and Numerical Method

3.1. Model Description

The cellular automaton model for nucleation and growth simulation consists of (1) the geometry of a cell; (2) the state of a cell; (3) the neighborhood configuration; and (4) several transition rules that determine the state of a given cell during one time step. In this work, the computational domain is divided into a uniform orthogonal arrangement of cells and the selected neighborhood configuration contains the first eight nearest neighbors. Each cell is characterized by several variables (such as temperature, concentration, crystallographic orientation, solid fraction) and state (e.g., solid or liquid). The cellular automaton evolves in discrete time step, and the state of a cell at a particular time is calculated from the local rule.

In order to predict the evolution of primary phase during the controlled cooling process as described in the previous section, the computational domain size was chosen to be \(1.2\,\text{mm} \times 1\,\text{mm}\), which is considered to be available for comparison with the microstructures observed experimentally. The whole domain was divided into 400\times 330 cells with a cell size of 3\,\mu m. As the thermal diffusivity of metallic alloys is 3–4 orders of magnitudes greater than the solute diffusivity, the kinetics for structure evolution can be assumed to be solute diffusion-controlled and the thermal diffusion can be considered to be complete in the microscopic scale. Thus, the temperature field inside the computational domain was assumed uniform and cooled down with a specified cooling rate. The governing equations and numerical algorithms for calculating nucleation, growth and solute field are described in details below.

3.2. Nucleation

For describing the heterogeneous nucleation in the bulk liquid, the continuous nucleation model is adopted in which a Gaussian distribution of nuclei is considered. The nuclei density increase \(d_n\) is induced by an increase in the undercooling \(dT\) according to the following Gaussian distribution\textsuperscript{[15]}:

\[
\frac{d_n}{d(T)} = \frac{n_{\text{max}}}{\sqrt{2\pi} \Delta T_d} \exp \left[-\frac{1}{2} \left(\frac{T - \Delta T_{\text{min}}}{\Delta T_d}\right)^2\right] \quad \text{....(1)}
\]

where \(\Delta T_{\text{min}}\) is the mean nucleation undercooling, \(\Delta T_d\) is the standard deviation, and \(n_{\text{max}}\) is the maximum density of nuclei given by the integral of this distribution from 0 to \(\infty\).

Thus the density of grains, \(n\), \((dT)\), formed at any undercooling \(\Delta T\), is given by

\[
n(dT) = \int_0^\infty d(n) d\Delta T' \quad \text{.........(2)}
\]

The nucleation parameters in Eq. (1), \(n_{\text{max}}, \Delta T_{\text{min}}\) and \(\Delta T_d\), are dependent upon the alloy composition and process variables. Generally, it is considered that the most important factor is the pouring temperature of a melt. The well-known free chill crystal theory\textsuperscript{[16]} and the separation nucleation theory\textsuperscript{[17]} suggest that a lot of crystals nucleate on the mold wall as a result of thermal undercooling immediately after pouring. These crystals grow out from the wall having a necked shape owing to the solute segregation, and subsequently are detached through mechanical breakage or thermal fluctuations in the liquid. It is apparent that lower pouring temperatures increase the number of free crystals that are able to survive and grow. However, in case of higher pouring temperatures, the crystals are expected to become smaller, and even disappear because of remelting. It is well known that the equilibrium liquidus of a crystal will decrease with a decrease of crystal size. This means that the

Fig. 1. Schematic diagram of experimental apparatus.
nucleation undercooling, which is needed for a nucleus becoming active for growth, increases as the crystal size decreases. Moreover, there is likely to be a larger variation in the potency of potential nuclei available under a higher melt superheat because of the thermal fluctuation. Such crystals have a wider potency distribution compared to the potency distribution of the crystals of a lower superheat. Accordingly, the different conditions for nucleus generation were implemented. The nucleation density increases, whereas the amount of undercooling and the standard deviation decrease with the decrease of the pouring temperature. In case of inoculation, it is obvious that not only the nucleus potency becomes very small and the amount of undercooling required to activate these nuclei is also small. The nucleation parameters with various pouring temperatures and inoculation used for the simulation are listed in Table 1.

At the beginning of simulation, nucleation undercoolings randomly generated according to the Gaussian distribution given by Eq. (1) are attributed to randomly chosen cells. The corresponding undercooling, ΔT_{nuc}, of these predetermined “nucleation” cells are stored. If a cell is chosen several times, only the smallest nucleation undercooling of the cell will be used. During a loop of calculation, if a predetermined nucleation cell is still liquid and the local undercooling, ΔT(t), is larger than the prescribed nucleation undercooling, ΔT_{nuc}, this cell becomes solid and its corresponding index is randomly defined to be an integer among 48 classes, which represents its crystallographic orientation.

3.3. Growth Kinetics and Orientation

In the present simulation, the growth velocity of a dendrite tip at a certain undercooling in the melt is calculated with the aid of KGT (Kurz–Giovanola–Trivedi) model.\(^{(18)}\)

The total undercooling in the solid/liquid interface, ΔT, is considered to be the sum of four contributions:\(^{(15)}\)

$$ΔT=ΔT_c + ΔT_p + ΔT_r + ΔT_k$$ ........................(3)

where ΔT_c, ΔT_p, ΔT_r, and ΔT_k are the undercooling contributions associated with concentration, temperature, curvature and attachment kinetics, respectively. The last term in the right hand of Eq. (3), the kinetic undercooling, becomes significant only at very high solidification velocities such as rapid solidification and is negligible under normal solidification conditions.\(^{(15)}\) In the present model, three contributions (solute, thermal and curvature effects) are taken into account. Therefore, the local undercooling of S/L interface at time \(t\), ΔT(\(t\)), is given by

$$ΔT(t)=T_l+[C(t)−C_0]·m−T(t)+ℓk(ℓ(t))$$ ........................(4)

where \(T_l\) is the equilibrium liquidus of the alloy, \(m\) is the liquidus slope of the phase diagram, \(C_0\) is the initial concentration of the alloy, and \(T\) is the Gibbs–Thomson coefficient. \(k(ℓ(t))\) is the mean curvature, the concentration and the temperature of the S/L interface at time \(t\), respectively.

According to the KGT model and an iteration method, the relationship between the growth velocity, \(v[ΔT(t)]\), and the local undercooling, ΔT(\(t\)), can be expressed as

$$v[ΔT(t)]=k_1·ΔT(t)^2+k_2·ΔT(t)^3$$ ........................(5)

where \(k_1\) and \(k_2\) are the coefficients: \(k_1=2.9·10^{-6}\) m/s K\(^{-2}\) and \(k_2=1.49·10^{-6}\) m/s K\(^{-3}\), for an Al–7mass\%Si alloy, respectively.

Once a cell has nucleated, it will grow with a preferential direction corresponding to its crystallographic orientation. The growth velocity is determined by the local undercooling ΔT(\(t\)) and calculated according to Eq. (5). The growth length, \(l(\(t\))\), is therefore given by

$$l(\(t\))=\sum_{\text{all}}\left[\frac{v[ΔT(t)]×Δt}{(\cos θ+|\sin θ|)}\right]$$ ........................(6)

where \(Δt\) is the time step, \(θ\) is the angle of the preferential growth direction of a solid cell with respect to the linking line between this solid cell and its liquid neighbor cell, and \(N\) indicates the iteration number. Then, the solid fraction of a given cell at a certain time, \(f_s(\(t\))\), can be expressed by

$$f_s(\(t\))=\frac{l(\(t\))}{L}$$ ........................(7)

where \(L\) is the spacing of the CA network. When \(f_s(\(t\))=1\), which means the growth front of the solid cell touches the center of its neighboring liquid cell, the neighboring cell transforms its state from liquid to solid and gets the same orientation index as this solid cell. The details of the growth algorithm have been given in the literature.\(^{(14)}\)

3.4. Solute Redistribution

It is assumed that the solid and liquid compositions at the solid/liquid interface are in equilibrium. When the solidification at the S/L interface occurs, the solute partition between liquid and solid is:

Table 1. Nucleation parameters used in the calculation.

| $T_{puse}$ (°C) | Inoculation | $n_{max}$ (m$^{-3}$) | Δ$T_{nuc}$ (°C) | Δ$T_s$ (°C) |
|-----------------|-------------|---------------------|-----------------|----------|
| 620             | no          | 7.2×10$^{11}$       | 0.5             | 0.1      |
| 625             | yes         | 1.0×10$^{12}$       | 0.5             | 0.1      |
| 630             | no          | 6.0×10$^{12}$       | 1.0             | 0.5      |
| 640             | no          | 3.8×10$^{11}$       | 2.0             | 1.0      |
| 660             | no          | 2.0×10$^{11}$       | 4.0             | 2.0      |
| 660             | no          | 9.0×10$^{10}$       | 8.0             | 4.0      |

Note: the relationship between the nucleus density for 3D, $n_{max}$ (m$^{-3}$), and the one for 2D, $n_{max}$ (m$^{-2}$), is given by $n_{max} = \frac{π}{6}(n_{max})^{3/2}$

© 2001 ISIJ
where $k$ is the partition coefficient, $C_e^s$ and $C_l^s$ are the interface equilibrium concentration in solid and liquid phases, respectively.

When a cell transforms its state from liquid to solid by nucleation or growth, its concentration will be changed according to Eq. (8). Consequently, this cell will liberate the amount of solute, $dC = C^s_l - kC^s_l$, which is assumed to be distributed to its liquid neighbor cells. As the solidification proceeds, the solidified cells reject solute to its neighbor liquid cells. Diffusion within the entire domain is then simulated. The governing equation for the solute redistribution in the liquid region is given by

$$\frac{\partial C_i}{\partial t} = \frac{\partial}{\partial x} \left[ D_i \frac{\partial C_i}{\partial x} \right] + \frac{\partial}{\partial y} \left[ D_i \frac{\partial C_i}{\partial y} \right] + C_i (1 - k) \frac{\partial f_i}{\partial t} \ldots \ldots (9)$$

where $t$ is time, $D_i$ is the solute diffusion coefficient in liquid, $f_i$ is the solid fraction, $k$ is the partition coefficient, and $x$ and $y$ are the coordinates system, respectively. The last term on the right hand side of the Eq. (9) indicates the amount of solute rejected at the solid/liquid interface.

The governing equation for diffusion in solid is given by

$$\frac{\partial C_s}{\partial t} = \frac{\partial}{\partial x} \left[ D_s \frac{\partial C_s}{\partial x} \right] + \frac{\partial}{\partial y} \left[ D_s \frac{\partial C_s}{\partial y} \right] \ldots \ldots (10)$$

where $D_s$ is the solute diffusion coefficient in solid. An explicit finite difference scheme was applied for calculating the solute diffusion in both the solid and liquid phases, and the zero-flux boundary conditions were used for the cells located at the surface of the computational domain. The thermal and physical properties used in the present calculation are listed in Table 2.

### Table 2. Thermal and physical properties used in the calculation.

| Property     | Value   |
|--------------|---------|
| $T_m$        | 660.0   |
| $T_{ext}$    | 577.0   |
| $C_{ext}$    | 10.77   |
| $k$          | 0.117   |
| $m$          | -6.0    |
| $\Gamma$     | 9.0 x 10^6 |
| $D_l$        | 1.0 x 10^9 |
| $D_s$        | 1.0 x 10^12 |

3.5. Calculation of the Interface Curvature

The interface curvature in a cell with the solid fraction $f_i$ is calculated from the following expression

$$K(A) = \frac{1}{dx} \left[ \frac{f_i(A) + \sum_{i=1}^{n} f_i(i)}{1 - 2 \frac{f_i(i)}{n+1}} \right] \ldots \ldots (11)$$

where $dx$ is the cell size, and $n$ is the number of neighboring cells which includes the first layer surrounding neighbor cells, $n=8$. The values of the curvature calculated using Eq. (11) vary from $1/dx$ to 0 for convex surfaces, and from 0 to $2/(dx)$ for concave surfaces.

3.6. Determination of Time Step

The stable time step for the MCA calculation is determined by

$$\delta t_{MCA} = 1 \min \left[ \frac{dx}{V_{max}} \frac{dx^2}{D_l} \right] \ldots \ldots (12)$$

where $V_{max}$ is the maximum growth velocity obtained by scanning the growth velocities of all interface liquid cells during each time step and $D_l$ is the solute diffusion coefficient in liquid. The flowchart of the present calculation is shown in Fig. 2.

4. Results and Discussion

4.1. Effects of Pouring Temperature

Figure 3 shows the predicted and experimental microstructures of an Al–7mass%Si alloy with a cooling rate of 0.3°C/sec for various pouring temperatures: (a) 620°C, (b) 630°C, (c) 640°C, and (d) 660°C. The figures on the upper row indicate the simulated microstructures and the lower ones the experimental results. The white phase in the figures indicates the primary $\alpha$-Al phase and the dark continuous matrix is the eutectic. It can be noted that under a low pouring temperature of 620°C, the primary $\alpha$ phase exhibits the globular morphology as shown in Fig. 3(a). However, with the increase of the pouring temperature, the primary phase transits gradually to an intermediate rosette-
like morphology with short stout arms as shown in Fig. 3(b) and 3(c), and finally transits to dendritic structures as shown in Fig. 3(d). These phenomena are considered to be due to different nucleation conditions with various pouring temperatures. The reported observations in a casting of ammonium chloride-water showed that for a pouring temperature close to the saturation temperature of a given solution, crystals appeared throughout the liquid during pouring. It can be noted that a low pouring temperature of 620°C, which is close to the liquidus, the nucleation occurs rapidly while the liquid is still in motion during pouring. Thus, the remelting of necked shape crystals and the fracturing of crystals can all contribute to the formation of new crystals, which promotes the formation of abundant crystal seeds in the molten alloy. Under this situation, crystals can grow simultaneously throughout the liquid region because the temperature field in the liquid is uniformly undercooled. The distribution of crystals and the spacing among them will be uniform and short because of a large number of nuclei generated simultaneously, leading to the suppression of dendritic growth in the melt. The coarsening of primary dendrite arms is achieved by slow and uniform cooling, and thus the primary dendrites become spherical.

In general, heterogeneous nucleation may not occur uniformly in the melt, resulting in non-uniform distribution of nuclei, i.e. non-uniform spacing between nuclei. This phenomenon can be found in both simulation and experiment, as shown in Fig. 3. Figure 4 describes the effect of the spacing between nuclei on the solute concentration field and the primary phase morphology with a pouring temperature of 625°C and a cooling rate of 0.3°C/sec at different solidification stages: (a) 1 min and (b) 3 min after start of solidification.
centration profiles in the liquid and the increase of darkness indicates the increase of solute concentration. It can be noted from Fig. 4 that the area labeled “A”, shows a larger spacing between nuclei than other area. During solidification, the solute concentration in the area “A” is lower than that in other area, as shown in Fig. 4 (a). According to Eq. (4), lower local concentration will result in higher local undercooling even in the same temperature field. Therefore, nuclei will grow faster in the direction toward the center of the area “A”, leading to the interface instability and the side branching of secondary arms. It can be seen from Fig. 4(b) that the primary phase in the area “A” exhibits the rosette-like features, instead of the globular structure as in the other area.

In case of lower cooling rates, both the local temperature and solute distributions in the liquid around the solid/liquid interface become uniform compared to higher cooling rates, giving rise to the interface stability. Furthermore, the effect of thermal undercooling in Eq. (4) decreases and the effect of curvature undercooling becomes more important. It is also known that the curvature effect enhances the spherical growth trend of crystals.

4.2. Effects of Cooling Rate

It is well known that the evolution of primary phase in solidification of alloys is closely related not only to the pouring temperature, but also to the cooling rate. In order to investigate the effects of cooling rate on the morphology of the primary phase evolution, the microstructures of an Al–7mass%Si alloy solidified at a pouring temperature of 620°C with various cooling rates were observed and compared with the simulation. Figure 5 indicates the obtained microstructures for various cooling rates: (a) 0.3°C/sec, (b) 3°C/sec, (c) 10°C/sec, and (d) 30°C/sec. The pictures on the upper row indicate the predicted microstructures and the ones on the bottom the experimental results. It can be clearly seen from the figures that with the increase of cooling rate, the primary phase morphology changes gradually from globular to rosette-like, and finally to dendritic structure.

Figure 6 indicates the solute concentration profiles in the liquid during the evolution of primary phase for various cooling rates. It can be seen that under the higher cooling rates, the solute is rapidly enriched in the liquid phase near the S/L interface because of the solute partition in the interface and little time available for the solute diffusion. Therefore, the interface stability is quickly destroyed, leading to the dendritic growth.

4.3. Effects of inoculation

In order to further demonstrate the mechanism of the globular structure formation, the effect of inoculation was also investigated by means of both simulation and experiment. Figure 7 shows the simulated and experimental microstructures of an Al–7mass%Si alloy with a pouring temperature of 625°C and a cooling rate of 0.3°C/sec: (a) with inoculation and (b) no inoculation. The pictures on the upper row indicate the simulated microstructures and the ones on the bottom the experimental results. It is interesting to note that in case of inoculation, the primary crystals exhibit more perfect and finer globular morphology with uniform distribution, compared to the case of without inoculation. This phenomenon can be easily understood in terms of an increased number of potential nuclei by the addition of inoculants. The previously reported experimental observation showed that heterogeneous substrate particles disperse finely in Al alloys, leading to the uniform concentration profiles in the liquid around the solid/liquid interface. The spacing between such particles can be within 10 μm. As the previous analysis, when the spacing between nuclei becomes shorter, the dendritic growth is suppressed.
Therefore, the inoculation will effectively benefit the formation of globular structure.

5. Conclusion

The evolution of globular and dendritic microstructures in solidification of an Al–7mass%Si alloy has been investigated by means of a modified cellular automaton (MCA) model. The present MCA model includes the nucleation and growth kinetics, the preferred growth orientation of a dendrite tip, as well as the constitutional undercooling and curvature undercooling. The solute redistribution including diffusion in solid and liquid is solved by the finite volume method. The morphology of the primary phase evolution, i.e., dendritic, intermediate rosette-like and globular, can be predicted by the MCA model. Results show that the globular structure can be obtained with lower pouring temperatures close to the liquidus combined with lower cooling rates. However, the dendritic structures can form when the pouring temperature is high even though the cooling rate is very low, and vice versa. It is also to be noted that inoculation enhances the formation of globular structure.

Acknowledgment

A part of this work was supported by the Brain Korea 21 Project. One of the authors (M. F. Zhu) is also grateful to KOSEF (Korea Science and Engineering Foundation) for the financial support by a Brain Pool system.

REFERENCES

1) M. C. Flemings: Metall. Trans., 22B (1991), No. 3, 269.
2) G. Hirt, R. Cremer, A. Winkelmann, T. Witulski and M. Zillgen: J. Mater. Process. Technol., 45 (1994), 359.
3) H. Lehuy, J. Massonneau, and J. Blain: J. Mater. Sci., 30 (1985), 105.
4) C. Vives: Metall. Trans., 23B (1992), 189.
5) H. I. Lee, J. I. Lee and M. I. Kim: Proc. of the 3rd Int. Conf. on Semi-solid Processing of Alloys and Composites, ed. by M. Kiuchi, The University of Tokyo, Tokyo, (1994), 281.
6) L. Salvo, M. Suery, Y. De Charentenay and W. Loue: Proc. of the 4th Int. Conf. on Semi-solid Processing of Alloys and Composites, eds. by D. H. Kirkwood and P. Kapranos, University of Sheffield, Sheffield, (1996), 10.
7) D. S. Shin, J. I. Lee, J. C. Lee and H. I. Lee: Proc. of the 4th Int. Conf. on Semi-solid Processing of Alloys and Composites, eds. by D. H. Kirkwood and P. Kapranos, University of Sheffield, Sheffield, (1996), 92.
8) S. G. R. Brown and J. A. Spittle: Mater. Sci. Technol., 5 (1989), No. 4, 362.
9) M. Kappaz and Ch.-A. Gandin: Acta Metall., 41 (1993), 345.
10) K. Y. Lee and C. P. Hong: ISIJ Int., 35 (1995), 637.
11) I. S. Cho and C. P. Hong: ISIJ Int., 37 (1997), 38.
12) Y. H. Chang, S. M. Lee, K. Y. Lee and C. P. Hong: ISIJ Int., 38 (1998), 97.
13) S. Y. Lee, S.M. Lee and C. P. Hong: ISIJ Int., 40 (2000), 48.
14) M. F. Zhu and C. P. Hong: ISIJ Int., 41 (2001), 436.
15) Ph. Thevoz: Sc. Dr. Thesis No. 765, Swiss Federal Inst. Tech., Lausanne, (1988).
16) B. Chalmers: J. Aust. Inst. Met., 8 (1963), 255.
17) A. Ohno, T. Motegi and H. Soda: Trans. Iron Steel Inst. Jpn., 11 (1971), 18.
18) W. Kurz, B. Giovanola and R. Trivedi: Acta Metall., 34 (1986), 823.
19) K. A. Jackson, J. D. Hunt, D. R. Uhllmann and T. P. Seward: Trans. Metall. Soc. AIME, 236 (1966), 149.
20) M. Kohashi, H. Tsukahara and T. Choh: J. Jpn. Inst. Light Met., 46 (1996), 73.
21) R. Sasikumar and R. Sreenivasan, Acta Metall., 42 (1994), 2381.
22) L. Nastac: Proc. of Modeling of Casting and Solidification Processes IV, ed. by C. P. Hong, J. K. Choi and D. H. Kim, Hanrimwon, Seoul, (2000), 31.